

TRANSACTIONS
of the
American
Foundrymen's Association



Proceedings of the
Thirty-seventh Annual Meeting

Chicago, Ill.

June 20 to 23, 1933

VOLUME XLI

EDITED BY ROBERT E. KENNEDY

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Summary of Proceedings of the 37th Annual Meeting

THE Thirty-Seventh Annual Convention of the American Foundrymen's Association was held June 20-23 at the Stevens Hotel, Chicago. An exposition of foundry equipment and supplies also was held in connection with this meeting, in the Exposition Hall of the hotel.

In addition to the regularly scheduled sessions of a technical character, the week was notable because of conferences held for discussion of the National Industrial Recovery Act, signed by President Roosevelt shortly before the convention opened. Four of these conferences were held under auspices of the Steel Founders' Society of America, Gray Iron Institute, Malleable Iron Research Institute and Foundry Equipment Manufacturers' Association, while the fifth, for nonferrous foundrymen, was held at the call of the A.F.A. Secretary. The nonferrous group at this meeting formulated an organization known as the Nonferrous Foundry Association for Industrial Recovery.

Of the regular association meetings, five were held jointly with other organizations, as follows: (a) Joint luncheon of A.F.A. and Foundry Equipment Manufacturers' Association; (b) Foundry Housekeeping Conference, with National Founders' Association; (c) Electric Power Costs Conference, with Steel Founders' Society of America, Gray Iron Institute and Malleable Iron Research Institute; (d) Materials Handling session, with Materials Handling Division of American Society of Mechanical Engineers; (e) Symposium on Cast Iron Tests and Specifications, with American Society for Testing Materials.

No evening meetings were held during the four days of the Convention, and the usual annual banquet also was dispensed with. Elimination of these events was occasioned by the desire to allow visiting members a maximum amount of time for visiting the Century of Progress Exposition.

OPENING SESSION

Tuesday, June 20, 10:00 A. M.

Presiding: President T. S. Hammond, Vice-President Frank J. Lanahan.

President Hammond opened the meeting with a short address of welcome and a review of Association activities for the previous year, after which he called upon Vice-President Lanahan, who responded briefly. Secretary C. E. Hoyt then discussed convention activities.

D. M. Avey, chairman of the International Relations Committee, proposed, inasmuch as the Institute of British Foundrymen was meeting at the same time, that the following cablegram be sent in the form of a resolution:

"The American Foundrymen's Association, at opening session Chicago Convention, conveys to Institute of British Foundrymen friendly greetings. May our several activities prosper and benefit the foundry world and our cordial relationships continue a pattern for molding international amity."

This cablegram was approved and dispatched, and in response the following letter was received subsequent to the convention:

INSTITUTE OF BRITISH FOUNDRYMEN

*St. Johns Street Chambers,
Deansgate, Manchester.
27th June, 1933*

T. S. Hammond, President
American Foundrymen's Association
Dear Mr. Hammond:

I am instructed by the Annual General Meeting of the Institute of British Foundrymen to thank you for your cablegram received at our Annual Conference in Cardiff on 21st June. Our members were very delighted to receive your cablegram and appreciate the cordial and friendly message which you sent us.

We trust that your conference in Chicago was as big a success as your previous conferences have been, and that you were able to see unmistakable signs of returning world prosperity.

You will be glad to know that our own Cardiff conference was a very happy and successful one, and we are all hoping that before long there will be a return of the trade which we all so much desire.

With kind regards, yours sincerely,

INSTITUTE OF BRITISH FOUNDRYMEN
TOM MAKEMSON, *General Secretary.*

President Hammond announced that next on the agenda was the appointment of a committee of three to select four members of the 1934 Nominating Committee, and he called upon Secretary Hoyt to make a statement regarding this procedure. The Secretary quoted Article IX, Section 1 of the By-Laws, reading as follows:

"There shall be created annually a Nominating Committee consisting of the last three living past-presidents of the Association and four other members to be elected by the members of the Association at least four months prior to the first day of the Annual Convention. The manner of their election shall be prescribed by the Board of Directors."

Secretary Hoyt then added the following statement:

"At a meeting of the Executive Committee held November 17, 1932, it was voted to recommend to the Board of Directors that

four members of the 1934 Nominating Committee, as provided for in Article IX, Section 1 of the By-Laws, be elected by the members at the 1933 annual meeting of the Association. This recommendation having been approved by the Board, it will be in order at the annual business meeting on Thursday to elect four members of the 1934 Nominating Committee, and the agenda of the present meeting provide for appointment by the President of a nominating committee of three to present nominations at the annual meeting."

The President then appointed the following: P. T. Bancroft, John Deere Harvester Works, East Moline, Ill.; John W. Bolton, The Lunkheimer Co., Cincinnati; J. H. Lansing, Danville Malleable Iron Div., Allith-Proutry Co., Danville, Ill. This committee was instructed to report at the Annual Business Meeting on Thursday, June 22.

The Opening Session was then adjourned.

CONFERENCE ON ELECTRIC POWER COSTS

Monday, June 19, 2:00 P. M.

(Held under auspices of a joint committee of the Steel Founders' Society of America, Gray Iron Institute and Malleable Iron Research Institute.)

Chairman, J. R. Allan, International Harvester Co., Chicago.

This conference was called by a joint committee to discuss data secured from a questionnaire sent to users of electric furnaces for melting metal for castings. Chairman Allan opened the meeting by presenting a summary of the data thus received. Following discussion of this report, a motion was made as follows:

"It is moved that the Committee be thanked for its activity in securing the data presented and that the Committee be continued to assemble further data on the subject."

The motion was seconded and carried.

SAND CONTROL SHOP-OPERATION COURSE (Session 1)

Tuesday, June 20, 8:30 A. M.

Chairman, W. G. Reichert, Singer Mfg. Co., Elizabeth, N. J.

This session was devoted to discussion of the application of sand control to continuous and jobbing foundries.

A. V. Leun, Bethlehem Steel Co., Bethlehem Pa., led the discussion on *Essential Apparatus for Sand Control*.

R. F. Harrington, Hunt-Spiller Mfg. Corp., Boston, presented a discussion on *Application of Sand Control*.

GRAY IRON SHOP-OPERATION COURSE (Session 1)

Tuesday, June 20, 8:30 A. M.

Chairman, P. T. Bancroft, John Deere Harvester Works, East Moline, Ill.

This session was devoted to a discussion of *Small Cupolas and Their Operation*; discussion led by D. J. Reese, Whiting Corp., Harvey, Ill.

FOUNDRY SAND RESEARCH

Tuesday, June 20, 10:30 A. M.

Chairman, R. F. Harrington, Hunt-Spiller Mfg. Corp., Boston.

The following papers were read and discussed:

Grain Structure Effect on Mold Permeability Control, by H. W. Dietert and Frank Valtier, H. W. Dietert Co., Detroit. (Read by Mr. Dietert.)

Effect of Silt on Bonding Strength of Clay, by Dr. H. Ries and R. C. Hills, Cornell University, Ithaca, N. Y. (Read by Dr. Ries.)

Sea Coal Effects on Sand, by Ben Hird, Keen & Nettlefolds, Ltd., Newport, Mon., England. The 1933 exchange paper of the Institute of British Foundrymen. (In absence of author, read by John Grennan, University of Michigan, Ann Arbor, Mich.)

Following presentation of these papers, Dr. Ries presented an informal report on the progress of activities of the Committee on Foundry Sand Research.

JOINT A.F.A. AND F.E.M.A. LUNCHEON
Tuesday, June 20, 12:30 P. M.

Presiding: T. S. Hammond, President, A.F.A.; H. S. Simpson, President F.E.M.A.

Following the luncheon, General Hammond as President of the A.F.A. made a few remarks and then turned the meeting over to Mr. Simpson as President of F.E.M.A. Mr. Simpson then introduced Fred Sargent, President of Chicago & Northwestern Ry., Chicago, who delivered an address on "Business and Politics."

FOUNDRY REFRACTORIES
Tuesday, June 20, 2:30 P. M.

Chairman, A. V. Leun, Bethlehem Steel Co., Bethlehem, Pa.

Two papers were presented and discussed, as follows:

Cupola Refractories, by Dr. E. E. Marbaker, Industrial Research & Engineering Co., Pittsburgh. (In absence of author, read by J. M. McKinley, North American Refractories Co., Cleveland.)

Factors Affecting the Service of Clay Refractories in Iron Foundries, by Robt. C. Zehm, The Stevenson Co., Wellsville, Ohio.

FOUNDRY COSTS
Tuesday, June 20, 2:30 P. M.

The Foundry Costs Committee developed a unique type of session by sponsoring a 4-act play, *Building for Profits*. The play, effectively dramatizing the benefits of an efficient cost system, was written by A. E. Grover, Cost Director of the Gray Iron Institute, and staged by Mr. Grover and the Cost Committee under direction of committee Chairman Jas. L. Wick, Jr., Falcon Bronze Co., Youngstown, Ohio. The cast included the following:

A. E. Grover, Gray Iron Institute, Cleveland.

Jas. L. Wick, Jr., Falcon Bronze Co., Youngstown, Ohio.

R. E. Belt, Malleable Iron Research Institute, Cleveland.

Morris Shirk, Berea, Ohio.

Geo. B. Michie, Electro-Refractories & Alloys Corp., Buffalo.

T. C. Watts, Falcon Bronze Co., Youngstown, Ohio.

T. W. Campbell, Electro-Refractories & Alloys Corp., Buffalo.

Luke U. Milward, Electro-Refractories & Alloys Corp., Buffalo.

Benj. D. Fuller, Whitehead Bros. Co., Cleveland.

John H. Diedrich, Blackhawk Foundry & Machine Co., Davenport, Iowa.

W. W. Kerlin, Gray Iron Institute, Cleveland.

SAND CONTROL SHOP-OPERATION COURSE (Session 2)

Wednesday, June 21, 8:30 A. M.

Chairman, Dr. H. Ries, Cornell University, Ithaca, N. Y.; *Vice-Chairman*, W. G. Reichert, Singer Mfg. Co., Elizabeth, N. J.

The subject discussed at this session was *Classification and Specification of Sands*. One paper was presented, as follows:

Application of Sand Testing by the Producer, by C. M. Hardy, Houghland & Hardy, Evansville, Ind.

GRAY IRON SHOP-OPERATION COURSE (Session 2)

Wednesday, June 21, 8:30 A. M.

Chairman, John Grennan, University of Michigan, Ann Arbor, Mich.

Discussion at this session was on *Gating Terms*, led by H. W. Dietert, U. S. Radiator Corp., Detroit.

CAST IRON—MELTING

Wednesday, June 21, 10:00 A. M.

Chairman, John W. Bolton, Lunkenheimer Co., Cincinnati; *Vice-Chairman*, Max Kuniansky, Lynchburg Foundry Co., Lynchburg, Va.

Papers read and discussed at this meeting were as follows:

Experience of a Gray Iron Jobbing Foundry with the Rocking-Type Electric Furnace, by C. R. Culling, Carondelet Foundry Co., St. Louis.

Air-Furnace Cast Iron, by D. P. Forbes, Gunite Corp., Rockford, Ill.

Effect of Slags on Cupola Operation and Metal Structure, by R. H. Bancroft and A. C. Myers, Perfect Circle Co., Newcastle, Ind. (Presented by Mr. Bancroft.)

ALLOYS IN STEEL CASTINGS

Wednesday, June 21, 10:00 A. M.

Chairman, R. A. Bull, Chicago; *Vice-Chairman*, W. C. Hamilton, American Steel Foundries, East Chicago, Ill.

Four papers reviewing the effects of alloys in steel castings were presented and discussed, as follows:

Vanadium in Steel Castings, by Jerome Strauss, Vanadium Corp. of America, Bridgeville, Pa.

Chromium in Steel Castings, by J. H. Critchett, Union Carbide & Carbon Research Laboratories, New York.

Molybdenum in Cast Steel, by H. W. Gillett and J. L. Gregg, Battelle Memorial Institute, Columbus, Ohio. (Presented by Dr. Gillett.)

Properties and Uses of Some Cast Nickel Alloy Steels by A. G. Zima, International Nickel Co., New York.

NONFERROUS CASTINGS AND BUSINESS MEETING
OF NONFERROUS DIVISION*Wednesday, June 21, 10:00 A. M.*

Chairman, Sam Tour, Lucius Pitkin, Inc., New York; *Vice-Chairman*, Jerome Strauss, Vanadium Corp. of America, Bridgeville, Pa.

Papers presented and discussed were as follows:

Effect of Small Additions of Aluminum, Silicon and Phosphorus to a Leaded Bronze, by H. J. Roast, McGill University, Montreal, Canada.

A Study of Six Bearing Metals, by O. E. Harder, Battelle Memorial Institute, Columbus, Ohio, and C. S. Cole, Copper and Brass Research Association, New York. (Presented by Mr. Cole.)

Beryllium-Copper Castings, by E. F. Cone, New York. (In absence of author, read by Mr. Riley of the Beryllium Development Corp., New York.)

The business meeting of the Nonferrous Division was held immediately following, Division Chairman Tour presiding.

Vice-Chairman Strauss announced the death of H. F. Seifert, who was chairman of the Committee on Recommendations for Design of Nonferrous Castings and a member of the division Advisory Committee. A resolution was presented and approved, to be presented in appropriate form to the family of Mr. Seifert and to the Westinghouse Electric & Mfg. Co., East Pittsburgh, Pa., with whom Mr. Seifert had been connected for a number of years.

Chairman Tour then reviewed the activities of the division for the past year. Committee reports were presented by H. M. St. John, Detroit Lubricator Co., Detroit, as chairman of the Round-Table Committee, and by R. L. Binney, Binney Castings Co., Toledo, Ohio, as chairman of the Program Committee. T. E. Kihlgren, International Nickel Co., Bayonne, N. J., chairman of the Symposium on Deoxidation and Degasification, reviewed the purpose of the symposium and stated that it was the object to carry this subject through future meetings.

MALLEABLE IRON

Wednesday, June 21, 10:00 A. M.

Chairman, Fred L. Wolf, Ohio Brass Co., Mansfield, Ohio.

Three papers were presented and discussed, as follows:

High-Strength Wear-Resistant Malleable Iron, by Rebecca Hall, Chicago. (In absence of author, presented in abstract by E. K. Smith, Electro Metallurgical Co., Chicago.)

Effect of Superheat on Annealing of Malleable Iron, by A. E. White and R. Schneidewind, University of Michigan, Ann Arbor, Mich. (Read by Dr. Schneidewind.)

Sand Control in a Continuous Molding System, by Charles Morrison, Saginaw Malleable Iron Div., General Motors Corp., Saginaw, Mich.

Chairman Wolf announced that the Honorary Degree of Doctor of Science had been conferred upon H. A. Schwartz, member of the Malleable Division Advisory Committee, by Rose Polytechnic Institute on June 10.

STEEL DIVISION ROUND-TABLE CONFERENCE

Wednesday, June 21, 12:30 P. M.

Chairman, P. E. McKinney, Bethlehem Steel Co., Bethlehem, Pa.; *Vice-Chairman*, F. A. Melmoth, Detroit Steel Castings Co., Detroit.

Following luncheon, the following paper and reports were presented and discussed:

Coreless Induction Furnaces, by W. E. McKibben, General Electric Co., Schenectady, N. Y. (In absence of author, read by Mr. Meyer, Ajax Electrothermic Corp., Philadelphia.)

Report of Chairman of Steel Division, presented by F. A. Melmoth, Detroit Steel Casting Co., Detroit, Chairman.

Report of Committee on Specifications for Steel Castings, presented by W. C. Hamilton, American Steel Foundries, East Chicago, Ind., committee chairman.

Report of Committee on Classification of Steels for Castings. (In absence of committee chairman, presented by R. A. Bull, Chicago.)

Report of A.F.A. Representative on Joint Committee on Investigation of Effect of Phosphorus and Sulphur in Steel, presented by R. A. Bull, Chicago, A.F.A. representative.

Following the presentation of these reports a motion was made that they be accepted as reports of progress. Motion put to vote and approved.

The report on Specifications for Steel Castings then was discussed in detail.

MALLEABLE DIVISION ROUND-TABLE CONFERENCE

Wednesday, June 21, 12:30 P. M.

Chairman, J. H. Lansing, Danville Malleable Iron Div., Allith-Prouty Co., Danville, Ill.

Following luncheon, three subjects were presented for discussion, subjects and discussion leaders being as follows:

Some Anomalies in Annealing; led by H. A. Schwartz, National Malleable & Steel Castings Co., Cleveland.

High-Temperature Properties of Malleable Iron; led by H. W. Maack, Crane Co., Chicago.

Cracks and Pouring Temperatures; led by E. M. Handley, Chain Belt Co., Milwaukee.

NONFERROUS DIVISION ROUND-TABLE CONFERENCE

Wednesday, June 21, 12:30 P. M.

Chairman, H. M. St. John, Detroit Lubricator Co., Detroit.

This conference was organized for informal discussion of subjects proposed informally by those present. Among the subjects thus discussed were laboratory equipment, and dirt and gas porosity.

SAND CONTROL SHOP-OPERATION COURSE (Session 3)

Thursday, June 22, 8:30 A. M.

Chairman, H. W. Dietert, U. S. Radiator Corp., Detroit.

This session was devoted to discussion of, first, defects in castings due to sand, and second, improvement of surface qualities. On the first topic, Chairman Dietert presented a chart outlining the various defects and remedies for them.

GRAY IRON SHOP-OPERATION COURSE (Session 3)

Thursday, June 22, 8:30 A. M.

Subjects discussed, and leaders of discussion at this session, were as follows:

Cast Iron Structures; led by E. K. Smith, Electro Metallurgical Co., Chicago.

Refractories for a Small Cupola; led by G. Olson, Illinois Malleable Iron Co., Chicago.

The Recommendations for Cupola Refractories, prepared under auspices of the Division of Simplified Practice, Bureau of Standards, were read and discussed. A motion was approved that they be referred back to the Joint Committee on Foundry Refractories for revision.

STEEL CASTINGS

Thursday, June 22, 10:00 A. M.

Chairman, F. A. Lorenz, Jr., American Steel Foundries, Chicago; *Vice-Chairman*, Geo. Batty, Steel Castings Development Bureau, Narberth, Pa.

The following papers were presented and discussed:

Studies on Solidification and Contraction and Their Relation to Formation of Hot Tears in Steel Castings, by C. W. Briggs and R. A. Gezelius, Naval Research Laboratory, Bellevue, Anacostia, D. C. (Read by Mr. Briggs.)

Dynamic Properties of Steel Castings as Determined by the Impact Test, by Fred Grotts, Continental Roll & Steel Foundry Co., East Chicago, Ind.

Advantages to Foundrymen of Radiographic Inspection, by H. R. Isenburger, St. John X-Ray Corp., New York.

CAST IRON

Thursday, June 22, 10:00 A. M.

Chairman, Jas. T. MacKenzie, American Cast Iron Pipe Co., Birmingham, Ala.; *Vice-Chairman*, R. S. MacPherran, Allis-Chalmers Mfg. Co., Milwaukee.

The following papers were presented and discussed:

Properties of Gray Cast Iron as Affected by Casting Conditions, by C. M. Saeger, Jr. and E. J. Ash, Bureau of Standards, Washington. (In absence of authors, presented in abstract by Jas. T. MacKenzie, American Cast Iron Pipe Co., Birmingham, Ala.)

Titanium in Cast Iron, by G. F. Comstock, Titanium Alloy Mfg. Co., Niagara Falls, N. Y. (In absence of author, presented in abstract by Mr. Welling.)

High-Silica Heat-Resisting Cast Irons, by A. L. Norbury and H. H. Morgan, British Cast Iron Research Association, Birmingham, England. (In absence of authors, read by J. S. Vanick, International Nickel Co., New York.)

Martensitic Quenching of Cast Irons, by Georges R. Delbart, Ateliers des Anciens Etablissements Cail, Denain, France. The 1933 exchange paper of the French Foundry Technical Association. (In absence of author, presented in abstract by H. Bornstein, Deere & Co., Moline, Ill.)

Silica and Manganese as Deoxidizers in Cast Iron, by A. H. Dierker, Ohio State University, Columbus, Ohio.

SYMPOSIUM ON DEOXIDATION AND DEGASIFICATION
OF BRONZE FOUNDRY ALLOYS

Thursday, June 22, 10:00 A. M.

Chairman, T. E. Kihlgren, International Nickel Co., Bayonne, N. J.; *Vice-Chairman*, Dr. C. H. Lorig, Battelle Memorial Institute, Columbus, Ohio.

This symposium was organized for discussion of the subject as related to three types of alloys. The three classes and the leaders of discussion on them, were as follows:

Alloy Group 1—Valve Bronzes, 85-5-5 type; discussion led by John W. Bolton, Lunkenheimer Co., Cincinnati.

Alloy Group 2—Tin Bronzes, 88-11 type; discussion led by O. W. Ellis, Ontario Research Foundation, Toronto, Canada. In presenting this discussion, Mr. Ellis read a paper on *Mechanism of Inverse Segregation*.

Alloy Group 3—Leaded Bearing Bronzes, 80-10-10 type; discussion led by G. H. Clamer, Ajax Metal Co., Philadelphia.

ANNUAL BUSINESS MEETING

Thursday, June 22, 1:30 P. M.

Report of the Annual Business Meeting will be found on page xv.

FOUNDRY HOUSEKEEPING CONFERENCE

Thursday, June 22, 2:30 P.M.

(Joint meeting of A.F.A. and National Founders' Association.)

Chairman, S. W. Utley, Detroit Steel Casting Co., Detroit, Past-President of A.F.A. and President of National Founders' Association.

This session was called for discussion of problems of foundry housekeeping, especially as related to the problem of dust control. The following papers were presented and discussed;

The Dust Problem in the Foundry Industry, by Dr. E. G. Meiter, Employers Mutual Liability Insurance Co., Milwaukee.

Legal Aspects and Employer Responsibility, by Alfred C. Hirth, Williams Eversman & Morgan, Toledo, Ohio.

Keeping a Clean and Orderly Foundry, by A. D. Lynch, Ohio Brass Co., Mansfield, Ohio.

SAND CONTROL SHOP-OPERATION COURSE (Session 4)

Friday, June 23, 8:30 A.M.

Chairman, Dr. H. Ries, Cornell University, Ithaca, N. Y.

This session was devoted to informal discussion of problems on sand control as brought forward by those present.

GRAY IRON SHOP-OPERATION COURSE (Session 4)

Friday, June 23, 8:30 A.M.

Chairman, H. Bornstein, Deere & Co., Moline, Ill.

The general subject under discussion was *Alloys in Cast Iron*. Subjects presented, and discussion leaders, were as follows:

Chromium; led by E. K. Smith, Electro Metallurgical Co., Chicago.

Nickel; led by J. S. Vanick, International Nickel Co., New York.

Molybdenum; led by E. R. Young, Climax Molybdenum Co., Detroit.

Vanadium; led by Jerome Strauss, Vanadium Corp. of America, Bridgeville, Pa.

CAST IRON TESTS AND SPECIFICATIONS

Friday, June 22, 10:00 A.M.

(Joint session of A.F.A. and American Society for Testing Materials.)

Chairman, H. Bornstein, Deere & Co., Moline, Ill., Chairman of A.F.A. Gray Iron Division and of A.S.T.M. Committee A-3 on Cast Iron.

The purpose of this meeting was to present practical information on tests and specifications, and the various phases of the subject were presented in a series of papers, as follows:

Introduction, by H. Bornstein, Deere & Co., Moline, Ill.

Significance of Testing Cast Iron and Limitations of Testing, by H. Bornstein.

Correlation of Test Bar and Casting, by R. S. MacPherran, Allis-Chalmers Mfg. Co., Milwaukee.

Transverse and Tensile Tests, Modulus of Elasticity, Fatigue and Impact Tests, by R. S. MacPherran.

A.S.T.M. Specification A48-32T, by John W. Bolton, Lunkenheimer Co., Cincinnati.

Mechanism of Testing, by Jas. T. MacKenzie, American Cast Iron Pipe Co., Birmingham, Ala.

Factors in Production of Test Bars, by Jas. T. MacKenzie.

Wear Tests and Value of Hardness Testing for Control of Product, by A. L. Boegehold, General Motors Research Laboratories, Detroit. (In absence of Mr. Boegehold, presented by R. F. Harrington, Hunt-Spiller Mfg. Corp., Boston.)

MATERIALS HANDLING

Friday, June 23, 10:00 A.M.

(Joint meeting of A.F.A. and Materials Handling Division of American Society of Mechanical Engineers.)

Chairman, Jervis B. Webb, Jervis B. Webb Co., Detroit.

Papers presented and discussed at this meeting were as follows:

The Field of Materials Handling in Small Foundries, by Max A. Amos, Standard Automotive Parts Co., and Robt. W. Grace, Muskegon, Mich. (Presented by Mr. Amos.)

The Field of Materials Handling in Semi-Production Foundries, by R. G. Wieland, Forest City Foundries Co., Cleveland.

APPRENTICE TRAINING CONFERENCE

Friday, June 23, 12:30 P.M.

Presiding, John H. Ploehn, French & Hecht, Inc., Davenport, Iowa.

This conference was for discussion of the present needs of the industry as related to apprentice training. A. B. Peirce, Director of Department of Industrial Education, National Metal Trades Association, Chicago, served as the conference leader.

CONFERENCE ON FOUNDRY EDUCATION IN ENGINEERING SCHOOLS

Friday, June 23, 2:00 P.M.

Chairman, Frank J. Lanahan, Fort Pitt Malleable Iron Co., Pittsburgh; *Vice-Chairman*, Frank G. Steinebach, *The Foundry*, Cleveland.

This conference was for the purpose of bringing together representatives of the various schools and executives of the foundry industry to discuss problems related to foundry instruction in engineering schools, and for discussion of a report presenting a recommended outline for a foundry course as submitted by the Joint Committee on Foundry Education in Engineering Schools. This recommendation was accepted.

Annual Business Meeting

AMERICAN FOUNDRYMEN'S ASSOCIATION

STEVENS HOTEL, CHICAGO, JUNE 22, 1933

Presiding, Frank J. Lanahan, Vice-President.

In opening the annual business meeting, Mr. Lanahan said he was sorry to announce that President Hammond had been called away and could not be at this meeting, and that it would be his prerogative, as Vice-President, to preside at this session.

The Chair called on Executive Secretary Hoyt, who stated that with the permission of the members, the annual reports of the Executive Secretary, Treasurer and Manager of Exhibits would be presented to the Board of Directors following the close of the fiscal year, July 1st.

He then expressed appreciation of the loyal support of those members who, through these trying times, had continued their support, and those exhibitors who, under very adverse conditions, came through so loyally and helped to make the convention a successful one. He also expressed appreciation to all authors and committee members who, through their contributions, had helped so splendidly during the year.

The Chair then called on Technical Secretary R. E. Kennedy, who presented an abstract of his report on technical activities.

D. M. Avey, Chairman of the Committee on International Relations, read two cablegrams of felicitations from the Institute of British FOUNDRYMEN and the French Foundry Technical Association. Mr. Avey also reported on plans being made for the Fifth World Foundry Congress which is scheduled to be held in the United States in 1934.

Secretary Hoyt then presented the report of the 1933 Nominating Committee, which was as follows:

"The 1933 Nominating Committee of A.F.A., elected in accordance with the provisions of the By-Laws, met in Cleveland, March 4, at which meeting the following were nominated for officers and directors:

For President, to serve one year—

Frank J. Lanahan, President, Fort Pitt Malleable Iron Co., Pittsburgh.

For Vice-President, to serve one year—

Dan M. Avey, Editor, *The Foundry*, Cleveland.

For Directors, each to serve three years—

George Batty, Technical Director, Steel Castings Development Bureau, Narberth, Pa.

T. S. Hammond, President, Whiting Corp., Harvey, Ill.

R. J. Teetor, General Manager, Cadillac Malleable Iron Co., Cadillac, Mich.

R. F. Harrington, Metallurgist, Hunt-Spiller Mfg. Corp., Boston.

Jas. L. Wick, Jr., President and General Manager, Falcon Bronze Co., Youngstown, Ohio.

For Director, to serve one year—

Dr. H. Ries, Department of Geology, Cornell University, Ithaca, N. Y., and Technical Director, Committee on Molding Sand Research.

"This report was submitted to the membership by mail on March 21. As no other candidates for officers and directors have been nominated, as provided for in the By-Laws, it will be in order at this Annual Meeting to declare the candidates named in the report of the committee duly elected to the offices for which they were nominated, to take office at the annual meeting of the Board of Directors following the close of this Convention."

Past-President E. H. Ballard, General Electric Co., West Lynn, Mass., then presented the following motion: "That these candidates be elected to the respective offices for which they were nominated in accordance with the By-Laws." The motion was seconded by Past-President A. B. Root, Jr. Secretary Hoyt put the motion, which was unanimously approved.

Vice-President Lanahan expressed appreciation for his election as President of the Association. He then introduced the newly elected Vice-President and Directors.

Secretary Hoyt reported that at the meeting of the Board of Directors, held Wednesday, June 21, a recommendation was approved to the effect that President Hammond be elected to Honorary membership on his retirement as president of the Association. Secretary Hoyt moved adoption of this recommendation, and on being seconded by Vice-President elect D. M. Avey, the motion was unanimously carried.

Secretary Hoyt then made the following statement:

"At the Opening Session on Monday, the Secretary reported that in accordance with provisions of the By-Laws, the Board of Directors had unanimously approved of the election of four members of the 1934 Nominating Committee at this annual meeting. Accordingly, President Hammond appointed a committee of three members, on nominations."

Chairman Lanahan then called for the report of this committee, which was read by P. T. Bancroft, as follows:

"Your committee begs leave to submit the names of the following members to act on the 1934 Nominating Committee:

W. C. Hamilton, American Steel Foundries, East Chicago, Ind.
E. E. Griest, Chicago Railway Equipment Co., Chicago.
A. E. Boegehold, General Motors Research Laboratories, Detroit.
E. F. Hess, Ohio Injector Co., Wadsworth, Ohio.

Respectfully submitted,

P. T. BANCROFT, *Chairman*
JOHN W. BOLTON
J. H. LANSING"

Past-President Ballard moved that the report of the committee be accepted, that nominations be closed, and that the meeting ballot on names presented; motion seconded and carried. A ballot was taken and the four members recommended were unanimously elected members of the 1934 Nominating Committee.

Chairman Lanahan next called on Past-President Ballard as chairman of the Board of Awards. Mr. Ballard responded as follows:

"Mr. President and Members of A.F.A., I have been delegated to perform a most pleasant duty in presenting to a very distinguished chemist and diligent worker in the Nonferrous Industry, a symbol of appreciation of his accomplishments, his untiring efforts, bringing into light that which remained in darkness to the industry

with which he has been associated for nearly forty years. That he is a searcher for truth is evidenced by the fact that he was a pioneer in applying chemistry in the Nonferrous Metals Industry.

"His early training in City Manual Training School of Philadelphia, and later at the University of Pennsylvania where he graduated with degree of B.S., followed by work in the private chemical laboratory of a noted mineral chemist, ideally fitted him for work—not only with his hands, but with his brain. His achievement in developing the scientific use of scrap based on analysis was a distinct contribution to the industry. His development of high-strength alloys for special requirements in ferrous and nonferrous metals was another outstanding contribution, as was his invention of a refining process in 1901, for which he was awarded the Elliot Cresson Gold Medal of the Franklin Institute.

"Under his direction his own company developed several types of melting furnaces. Particularly noteworthy is the electric induction furnace. He has to his lasting credit many other contributions which are of equal importance and are elsewhere recorded.

"It is an old saying, 'If you want to have a job well done, secure a busy man.' This saying is truly born out in the case of our distinguished member, who has found time in addition to his personal business duties to serve as President of the American Institute of Metals, President of the American Society for Testing Materials, President of our own Association, for many years a member of the Board of Managers of the Franklin Institute and Chairman of the Committee on Science and the Arts. He is a Charter Member of the American Electrochemical Society of Metallurgical Engineers and a member of many other technical societies at home and broad. Notwithstanding his long years of activity, he is still continuing his contributions of time, money and energy for the advancement of the industry.

"Dr. Clamer, it is indeed pleasing to your host of friends to learn of the degree of Doctor of Science recently conferred upon you by Ursinus College. I cannot adequately express what a sincere pleasure it is for me on behalf of the American Foundrymen's Association to present you with the Joseph S. Seaman Gold Medal and Certificate, which you have so honorably and creditably earned. May your life be spared for many years to come and may this medal ever be to you a reminder of the esteem and affection in which you are held by your associates, to which we add: 'Well done, good and faithful servant.'

Mr. Ballard then presented to Dr. Clamer the Joseph S. Seaman Gold Medal of the A.F.A. and the official Certificate of Award. Dr. Clamer responded as follows:

"Mr. Vice-President, President-elect, Mr. Ballard and gentlemen, I am surely glad that this Seaman Award is this year not made posthumously. I say that because, as you may recall, on three previous occasions this Award has been so made, the ones chosen for the Award having passed into the Great Beyond during that interval from the time the choice was made until the day of the actual award. Now I am mighty glad one can be alive to hear such nice things said of one, and I assure you it is much better than to have such words said over one's coffin.

"I cannot accuse my good friend Bert Ballard of flattery—he is too sincere for that. Nor can I accuse him of over-statement, because he is too honest for that. But I am reminded of a true story that was told to me by our old friend Jesse Jones, one of our medalists and a man who passed on to the Great Beyond before the medal was awarded. It was the story of a molder in the Westinghouse foundry.

"It appears that this molder had devised a very ingenious method of molding a particular pattern. There was a certain very prolific writer for one of our technical journals who was told about this accomplishment, and he went to the Westinghouse foundry to see this man and was given a few rough sketches and was told about the method in a very few words. Notes were taken, and in a few weeks there appeared in the technical press many pages describing the accomplishments of this molder. Jesse passed these pages on to the molder and he looked them over very carefully to see that it was really his own stuff and, after discovering that it was his own stuff, his only remark was: 'Well, I'll be darned!'

"Well, I feel very much that way just at the present time. I will call your attention to the fact that a man cannot work along one channel if he is at all diligent without accomplishing a great deal in the period of about 35 years.

"I also call to your attention that it is not possible to accomplish a great deal by one's self. In the early days as a scrub chemist and with a small laboratory, I did the work by myself. I worked in the foundry with its technical problems. But after while I had an assistant, and a little later I had another assistant, and finally about all I had to do was to direct. Now I have often wondered whether I didn't lose a whole lot by that change, by not actually being there and doing the work but directing others to do it instead.

"The accomplishments which your Board of Awards has seen fit to recognize by the award of this wonderful medal, which was founded by our good old friend 'Daddy' Seaman, are not the result of my efforts alone but of my efforts plus those who have worked with me. I therefore accept this honor, Mr. President, not for myself alone but for those who have worked with me. I consider it a great honor to have this medal, and I highly appreciate that the Board of Awards should have considered the work of such an order of excellence as to have warranted this award. I therefore accept it, not for myself alone, but for myself and my associates, for whom I will act as custodian of this medal."

There being no further business, the meeting stood adjourned.

Minutes of Board Meetings

The following pages include information on the minutes of the following meetings of the Board of Directors and the Executive Committee of the Board held during the calendar year, 1933:

June 22—Final meeting of the 1932-33 Board of Directors, held during the Chicago Convention.

June 22—First meeting of the 1933-34 Board of Directors, held during the Chicago Convention.

August 22—Meeting of the Executive Committee, held in Chicago.

December 9—Second meeting of the 1933-34 Board of Directors, held in Cleveland.

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Minutes of Annual Meeting 1932-33 Board of Directors

AMERICAN FOUNDRYMEN'S ASSOCIATION
STEVENS HOTEL, CHICAGO, JUNE 22, 1933

Vice-President Frank J. Lanahan presiding, in absence of President T. S. Hammond.

The following responded to roll call: Directors C. S. Anderson, Fred Erb, R. M. Maull, D. M. Avey, W. D. Moore, N. K. B. Patch, Fred L. Wolf, E. H. Ballard, H. Bornstein and Franklin G. Smith; Directors-elect R. F. Harrington, R. J. Teetor, Jas. L. Wick, Jr. and H. Ries; members of the Advisory Board, R. A. Bull, G. H. Clamer, Benj. D. Fuller, A. B. Root, Jr. and S. W. Utley.

This meeting followed the annual Alumni Dinner, which was attended by the above named and the following: Past-Presidents L. W. Olson and S. T. Johnston; past-Directors Verne E. Minich, T. W. Pangborn, H. S. Simpson, S. C. Vessy and Arnold Lenz.

On motion by Mr. Wolf, seconded by Mr. Borstein and carried, the reading of minutes of the annual meeting of the 1932-33 Board of Directors held at Detroit, May 5, 1932, was dispensed with.

Executive Secretary C. E. Hoyt reported a special meeting of the Board of Directors which had been held the previous day, June 21, at which time it was voted to recommend to the annual meeting of the

Association that retiring President T. S. Hammond be elected to Honorary Membership.

Before proceeding with the regular order of business, Executive Secretary Hoyt reported that it was the recommendation of the officers of the Association that only matters of routine be dispensed with at this meeting, and that an adjourned meeting be held at the call of the President some time after the end of the fiscal year, June 30, when the books were audited and the reports of officers for the past year were prepared and made available. He stated further that if this was agreeable, the reports of officers, the next order of business, would be brief. Brief reports were then made by the Executive Secretary, Treasurer, Technical Secretary and Manager of Exhibits.

Local Affiliation

The Executive Secretary reported a communication received from the Chicago Foundrymen's Club requesting that the Board of Directors take under consideration the question of amending the By-Laws so as to make provision for local chapters of A.F.A. The Secretary also reported that during the Convention the Secretary of the Detroit Foundrymen's Association had verbally reported that the Executive Committee of the Detroit Association had given the question of local chapters of A.F.A. consideration, and had instructed him to advise the officers of the A.F.A. of their interest in the development of some plan for creating local chapters.

The Secretary stated that the question of local sections had been given consideration on numerous occasions, and that early this year and prior to the officials having been advised that any local associations were interested, there had been an exchange of correspondence on the subject of developing some plan for district activities of A.F.A. which might possibly lead to the forming of local sections.

Upon motion by Mr. Wolf, seconded by Mr. Patch and carried, the matter of affiliation with local chapters, or the creating of some plan of district representation, was referred to the incoming Board for consideration.

The Secretary reported that at the annual meeting of the Association held the afternoon of this day, Frank J. Lanahan had been elected President and Dan M. Avey Vice-President, to serve terms of one year each; that George Batty, T. S. Hammond, R. F. Harrington, R. J. Teetor and Jas. L. Wick, Jr. had been elected Directors to serve terms of three years each; and that Dr. H. Ries had been elected to fill the vacancy on the Board caused by the election of Dan M. Avey as Vice-President—the above officers and directors to take office at the annual meeting of the Board of Directors following adjournment of this, the final meeting of the 1932-33 Board. It was moved and carried that the report of election of officers and directors be made a matter of record in the minutes of this meeting.

On motion by Mr. Ballard, seconded by Mr. Wolf and carried, the meeting stood adjourned.

Respectfully submitted,

C. E. Hoyt, *Executive Secretary*

Minutes of First Meeting 1933-34 Board of Directors

AMERICAN FOUNDRYMEN'S ASSOCIATION
STEVENS HOTEL, CHICAGO, JUNE 22, 1933

President Frank J. Lanahan presiding.

The following responded to roll call: President Frank J. Lanahan, Vice-President D. M. Avey; Directors W. D. Moore, N. K. B. Patch, H. Ries, Fred L. Wolf, E. H. Ballard, H. Bornstein, Franklin G. Smith, R. F. Harrington, R. J. Teetor and Jas. L. Wick, Jr.; Executive Secretary C. E. Hoyt and Technical Secretary R. E. Kennedy; and members of the Advisory Board, R. A. Bull, G. H. Clamer, Benj. D. Fuller, A. B. Root, Jr. and S. W. Utley.

The Chair announced that the first order of business would be the organizing of the Board by the election of an Executive Secretary, Treasurer, Technical Secretary, Manager of Exhibits, Assistant Secretary and four Directors to serve as members of the Executive Committee, and that he would appoint the following as members of a Nominating Committee to retire and prepare nominations for the above-named offices: R. F. Harrington, *Chairman*; Franklin G. Smith, and N. K. B. Patch.

Mr. Harrington, as chairman, submitted the following report:

"Your committee appointed to nominate officers of the Association and members of the Executive Committee, in accordance with the provisions of the By-Laws, submits the following report:

(1) We recommend, as provided in the By-Laws, that the offices of Executive Secretary and Treasurer be combined and the Chairman so moves.

(Motion seconded and carried.)

(2) For Executive Secretary-Treasurer, we nominate C. E. Hoyt.

(3) For Technical Secretary, we nominate R. E. Kennedy.

(4) For Manager of Exhibits, we nominate C. E. Hoyt.

(5) For Assistant-Secretary-Treasurer, we nominate Miss Jennie Reininga.

(6) For four members of the Executive Committee, to serve with the President, Vice-President and Executive Secretary, we nominate the following Directors: T. S. Hammond, H. Bornstein, Fred L. Wolf and Jas. L. Wick, Jr."

Mr. Harrington moved acceptance of the report, which was seconded by Mr. Smith and carried.

It was moved by Mr. Bornstein, seconded by Mr. Harrington and carried, that President Lanahan, Vice-President Avey and Past-President Hammond serve as the Finance Committee for the ensuing year, with the past-president as Chairman.

It was moved by Mr. Ballard and seconded by Mr. Bornstein that the usual resolutions for the withdrawal and disbursement of Association funds, required by the Harris Trust & Savings Bank, be approved

and certified to by the Executive Secretary and a Director not an officer, as required by the resolutions. Motion carried.

On motion duly seconded, bonds for the Treasurer and Assistant Treasurer were fixed at \$5,000.00 each, premiums to be paid by the Association.

It was moved by Mr. Avey, seconded by Mr. Wolf and carried, that salaries in effect for the year ending June 30, 1933, be continued until the next or adjourned meeting of the Board of Directors.

It was moved, seconded and carried that the resolutions in effect during the past year, re defraying the expenses of members in attendance at committee meetings, be continued in effect until further action is taken by the Board.

It was moved by Mr. Harrington, seconded by Mr. Bornstein, that the President be authorized to make all appointments for standing and special committees not provided for in the By-Laws or by special act of the Board. Motion carried.

The Executive Secretary made a report on invitations for the 1934 annual Convention, which would also be the occasion for an International Foundry Congress, following which it was moved by Mr. Ballard and seconded by Mr. Wolf that a committee be appointed by the President to consider the invitations and make their recommendations to the Board. Motion carried.

It was moved by Mr. Wolf, seconded by Mr. Avey, that the matter referred to the new Board by the old Board, re local association chapters, be referred to a committee to be appointed by the President for consideration and recommendations to the Board. Motion carried.

It was moved by Mr. Ballard, seconded by Mr. Bornstein and carried, that the meeting adjourn subject to call of the President.

Respectfully submitted,

C. E. HOYT, *Executive Secretary*

Minutes of Executive Committee Meeting 1933-34 Board of Directors

AMERICAN FOUNDRYMEN'S ASSOCIATION
UNION LEAGUE CLUB, CHICAGO, AUGUST 22, 1933

PRESIDENT F. J. LANAHAN presiding.

Roll Call

Meeting held at the Union League Club, Chicago, with the following present: Pres. Frank J. Lanahan, Vice-Pres. D. M. Avey, H. Bornstein, F. L. Wolf, J. L. Wick, Jr., and C. E. Hoyt, members of the Executive Committee; Technical Secretary R. E. Kennedy; Past Presidents R. A. Bull and S. T. Johnston.

Secretary Hoyt in submitting an agenda for the meeting, reviewed briefly an exchange of correspondence between members of the committee and with Past President R. A. Bull on the subject of amendments to the

By-Laws re. classes of membership and making provision for local sections and chapters.

Advertising Policies

The Secretary asked the privilege of making a reply to a letter from Mr. Bull under date of August 14, on the subject of advertising in A.F.A. publications, calling attention to the statement made by Mr. Bull that "From the standpoint of precedent, A.F.A. has followed a very unusual policy in respect to advertising." Mr. Hoyt outlined the factors that had determined advertising policies and then moved that the President be authorized to appoint a committee of three to consider the whole question of advertising policy and submit a report for Board consideration and action. The motion was seconded by Mr. Bornstein and unanimously carried.

Revision of By-Laws

Articles and Sections of present By-Laws and proposed amendments and additions thereto were considered at length, and following separate action on each, it was moved, seconded and carried that the Executive Secretary be instructed to prepare copies of a revised draft of By-Laws as approved by the Executive Committee, and submit them to all Directors prior to the next meeting of the Board.

Principles of Practice

Secretary Hoyt, calling attention to the draft of Principles of Practice adopted by the Board in 1924, recommended that a new draft be prepared, less negative in character, setting forth the purposes and scope of activities of the Association.

Mr. Avey moved that the President be authorized to appoint a committee to draft a new declaration of Principles of Practice so worded as to make the A.F.A. the recognized foundry organization for technical research for all branches of the metal castings industry, and as distinct from activities of trade association groups. The motion was seconded by Mr. Bornstein and carried.

Membership Campaign

Secretary Hoyt outlined the membership drive conducted in 1920 under the direction of General Chairman, Past President Alfred E. Howell, resulting in a total of 455 new members for year ending Dec. 31st, 1920.

The Secretary also reported the membership campaign conducted in 1925 under the chairmanship of Vice-Pres. A. B. Root, Jr., resulting in a total of 501 new members for that year.

President Lanahan, expressing approval of periodical campaigns of this character, said he thought it would be unwise to inaugurate such a campaign at a time when business in the foundry industry was at such a low level, and when all foundrymen were concerned with or disturbed by the multiplicity of trade association activities in connection with the National Industrial Recovery Act.

It was the consensus of opinion that the logical time for launching a membership campaign would be after amendments to the By-Laws had been approved and when the campaign could be coupled with the pub-

licity to be carried on in connection with the staging of the International Foundry Congress in the Fall of 1934.

1934 Convention

President Lanahan announced that at the Annual Board Meeting all invitations for the 1934 Convention had been referred to a committee to be named by the President, and that it was his desire that the Executive Committee as a whole serve as members of the committee on time and place of the next convention. Mr. Lanahan then called on Secretary Hoyt for a report on invitations received.

Secretary Hoyt in submitting a report, stated that correspondence and conferences had narrowed the field down to two central points, Cleveland and Philadelphia. He then outlined the proposals and agreements which had been submitted for the use of the public auditoriums in these two cities and reported the co-operation pledged by the convention Bureaus in the respective cities.

The Secretary stated further that there had been an exchange of official correspondence in regard to dates, and that it would appear to be the unanimous opinion that the most acceptable date would be after the middle of October, following the annual conventions for the year of European Foundry Associations co-operating in these international events.

Following discussion it was unanimously voted to instruct the Executive Secretary to submit to the Board of Directors for letter ballot the recommendation of the Executive Committee that the 1934 Convention, Exposition, and International Foundry Congress be held in Philadelphia during the month of October.

Secretary Hoyt moved that the Executive Committee recommend to the Board of Directors that the 1935 Annual Convention of the Association be staged without an exhibit, and that announcement of this action be coupled with an announcement of the time and place of the 1934 Convention and Exhibit. The motion was seconded and carried.

Technical Secretary's Report

President Lanahan announced that the remaining time would be devoted to a discussion of the Technical Secretary's report which had been submitted to all Directors and printed in full in *TRANSACTIONS* for August.

Following discussion the meeting adjourned to convene again at the call of the President.

Respectfully submitted,
C. E. HOYT, *Executive Secretary.*

Minutes of Adjourned Annual Board Meeting 1933-34 Board of Directors

AMERICAN FOUNDRYMEN'S ASSOCIATION
CLEVELAND HOTEL, CLEVELAND, OHIO, DECEMBER 9, 1933

The meeting was called to order at 10:15 by President Lanahan.

Roll Call

Present, President Frank J. Lanahan, Vice-President D. M. Avey, Directors N. K. B. Patch, H. Ries, A. G. Storie, E. H. Ballard, H. Bornstein, Franklin G. Smith, R. F. Harrington, J. L. Wick, Jr., F. L. Wolf, Executive Secretary C. E. Hoyt, and Technical Secretary R. E. Kennedy.

Following roll call the Secretary read communications from absent directors George Batty, S. B. Cuthbert, and R. J. Teeter, and reported on the absence of Directors W. D. Moore, T. S. Hammond, and David Evans. President Lanahan requested the Secretary to extend sympathy to directors absent on account of illness.

The Executive Secretary read a summary of the minutes of the meeting of the Board held on June 22, and on motion, duly seconded, the minutes of that meeting were approved without further reading.

The Executive Secretary read his report on Membership and Finances. On motion the Secretary's report was received and ordered filed.

Membership

President Lanahan, discussing the Secretary's report on Membership Campaigns conducted in 1920 under the chairmanship of Vice-President Alfred E. Howell, and the campaign in 1925 under the chairmanship of Vice President A. B. Root, Jr., stated that he concurred in the Secretary's recommendation that the Board organize a committee for membership campaign, but agreed with the Secretary and others that it would be wasted effort to conduct a campaign before there was improvement in business conditions.

Directors Harrington, Ballard, Bornstein, and Avey offered suggestions for increasing membership and all were in agreement that the Association offered too many benefits and privileges to non-members, and voiced the opinion that limiting these to members would increase membership.

The Secretary called attention to membership letters recently issued by other technical associations, and emphasized the value of having these letters written and signed by members not of the official paid staff of the Association.

Finances

Supplementing the auditor's report* for the fiscal year ending June 30, 1933, showing an operating loss of \$3,568.39, the Executive Secretary submitted a comparative statement showing receipts and disbursements for the first five months of the year beginning July 1, 1932, and the year beginning July 1, 1933, showing cash on hand December 1, 1932, of \$4,261.93, and on December 1, 1933, \$4,501.34.

* A copy of the Auditor's Balance Sheet and Statement of Cash Receipts and Disbursements can be found on pages xxx to xxxii inclusive.

He also submitted an itemized statement of expenses for three years totaling as follows:

	Total Expenses
Year ending December 31, 1930.....	\$96,226.28
Year ending June 30, 1932.....	51,043.39
Year ending June 30, 1933.....	38,315.32

The Secretary called attention to the salary, clerical and stenographic item in the comparative statements, totaling \$33,176.74 in 1930, \$28,269.21 in 1932, and \$19,717.86 for the year ending June 30, 1933, stating that since July 1st there had been further reductions, and that on December 1st, the Association was operating on a salary budget basis of \$16,280.00.

Reserve Fund

Reporting on the Reserve Fund, the Executive Secretary stated that during the year 1932-33 this fund had been reduced by the sale of \$3,000 U. S. 3½% Treasury Notes to offset the operating deficit of that year, and that there had been no interest defaults in the reserve fund securities having a par value of \$32,500.00.

Award Fund

On Award Fund investments the Executive Secretary reported that on July 1 there had been a default in interest payment on \$5,000 City of Detroit 4½% Sewer Bonds, and submitted a letter from the Trust Department of the Harris Trust & Savings Bank, Custodian of our Reserve and Award Fund securities, recommending that authorization be voted by the Board that the City of Detroit Bonds be turned over to the Bondholders' Refunding Committee, and that on receipt of this information he had conferred with other financial organizations and then made a report to the members of the Board of Awards who had approved, by letter ballot, of this procedure which would require a resolution of the Board.

The following resolution was unanimously adopted:

"RESOLVED: That the Executive Secretary-Treasurer, Charles E. Hoyt, be and he hereby is authorized and directed to deposit for this Association, Five Thousand (\$5,000) City of Detroit 4½% Sewer Bonds due 1957, under and pursuant to the terms of the Deposit Agreement dated June 6, 1933, and to execute such assignment and letters of transmittal as may be necessary to accomplish such deposit."

Unfinished Business

The Secretary reported that at the annual meeting of the Board held on June 22 it was voted that salaries in effect for the year ending June 30, 1932 be continued until the next adjourned meeting of the Board. President Lanahan stated that the Finance Committee was not quite ready to make its recommendations for salaries and that this would be taken up later.

Committee Traveling Expenses

It was unanimously voted to keep in effect the following resolution adopted at the 1932 Annual Meeting.

"RESOLVED: That the Treasurer be authorized to reimburse the traveling expenses of Directors and committee members for attendance at any regularly called Board or committee meetings, with the following exceptions:

"When meetings are held in conjunction with other committees or associations, the Treasurer is authorized to determine what portion of the expense of attending such meetings shall be paid by the Association.

"No expenses shall be paid to Directors or committee members for attendance at meetings held during the week of the annual Convention of the Association unless specially authorized."

1934 Convention

The Secretary reported that at the meeting of the Board on June 22, the question of time and place of the 1934 Convention was referred to a committee to be appointed by the President; that the President had named the members of the Executive Committee as this committee, and that at a meeting held in Chicago on August 22 it was unanimously voted to recommend to the Board of Directors that the 1934 Convention, Exposition and International Foundry Congress be held in Philadelphia during the month of October, and that the Executive Secretary be instructed to submit this recommendation to the Board of Directors for letter ballot.

The Secretary reported that following satisfactory negotiations for lease on the Philadelphia Exhibition and Convention Halls and Commercial Museum, the recommendations of the Executive Committee were submitted to the Board of Directors on September 14th, who by letter ballot unanimously approved of holding a convention and exposition in Philadelphia, October 22 to 26, 1934.

It was ordered that the letter ballot on convention date be made a matter of record, and that the President and Executive Secretary be authorized to sign rental agreement with the Board of Trustees of the Philadelphia Commercial Museum, Exhibition and Convention Halls.

Amendments to By-Laws

Secretary Hoyt stated that all directors had received report of the meeting of the Executive Committee held on August 22, recommending changes and additions to By-Laws, and that with the agenda would be found the latest draft of all proposed changes and additions together with copy of the present By-Laws.

President Lanahan announced that approval of revisions to By-Laws would be the next order of business, and if there were no objections they would be taken up and passed upon article by article.

When this had been done it was unanimously voted to instruct the Executive Secretary to have printed the complete by-laws as approved by the Board of Directors, and submit them to the entire membership for approval by letter ballot.*

Principles of Practice

Dan M. Avey, Chairman of the Committee to prepare a statement of Principles of Practice, which when adopted would take the place of Principles of Practice adopted by the Board in November, 1924, submitted report for his committee.

* The revised by-laws as approved by letter ballot vote of the members are to be found on pp. xxxii to xxxix inclusive.

On motion, duly seconded, the report of the committee was unanimously adopted, and ordered printed in *TRANSACTIONS*.**

Report of Committee on International Relations

Mr. Avey, Chairman, reported interview held that morning with Mr. George White of Thomas Cook and Son, at which time consideration had been given to offering a pre-convention tour for overseas delegates, that would include a visit to the World's Fair in Chicago; that this question would be taken up with our European representative, Mr. Vincent Delport. It was the thought of the committee that such a plan might very considerably increase the attendance of overseas foundrymen.

On motion, the report of the Chairman of the committee was accepted.

Technical Secretary's Report

The Technical Secretary's report was mailed to all board members on July 15th. Secretary Kennedy very briefly reviewed some of the items in this report stating that a number of directors had made written comments and he requested further written comments and suggestions which would be helpful to him and his committees.

Report on the Cast Metals Handbook

Secretary Kennedy in submitting a report on the Cast Metals Handbook said that the Steel, Malleable Iron, and Gray Iron sections were practically completed; that there was still considerably more work to do on the Nonferrous section, and that this work would be rushed with all possible speed.

Advertising in Cast Metals Handbook

Secretary Hoyt stated that the Committee on Advertising in the Cast Metals Handbook, of which he was chairman, did not have definite recommendations to make, but later in the evening other members of the committee, E. H. Ballard and H. Bornstein, discussed the subject with Board members and outlined a report to be submitted to the President and members of the Board.

Report of Finance Committee on Salaries

The Finance Committee consists of the President, the Vice-President, and the immediate Past President who shall be chairman. In the absence of the chairman, Director T. S. Hammond, Vice-President Avey submitted the report which stated that in the absence of Past President Hammond, Past Presidents Ballard and Patch had conferred with the two members of the committee, and concurred in the report of the committee, as follows:

Salary, Executive Secretary-Treasurer \$320.00 per month; no change.

Salary, Technical Secretary \$425.00 per month; recommend increase of \$25.00 per month effective January 1, 1934.

Salary, Manager of Exhibits \$320.00 per month; no change.

Salary, Assistant Secretary-Treasurer \$184.00 per month; recommend an increase of \$8.00 per month effective January 1, 1934.

** These principles of practice are to be found on pp. xxxix and xl.

Mr. Bornstein moved the acceptance of the report of the Finance Committee, with the understanding that the Executive Committee, acting on the recommendation of the Finance Committee would have authority to make any adjustment in salaries and compensations of employees during the balance of the fiscal year ending June 30, 1934.

Committee on Membership

Acting on the recommendations made in the report of the Executive Secretary on Membership, it was moved, seconded, and carried, that the President be authorized to appoint a committee on Membership to conduct a membership campaign to be inaugurated when in the opinion of the Executive Committee it seemed advisable to do so.

Exhibit Committee

It was moved, seconded and carried, that the President be given authorization for the appointment of an exhibit committee if it was found advisable.

Convention and Exhibit Advertising

Secretary Hoyt submitted a 1934 Calendar dummy, suggesting that a calendar might be used in place of the customary pocket calendar card. Authorization for publishing a calendar was voted.

Election to Life Membership

In recognition of long and honorable connection with the foundry industry, beginning in 1862, at the age of twelve, Mr. J. B. Goostray of South Boston, Mass., was elected a Life Member of the Association. Vice-President Dan M. Avey was authorized to announce Mr. Goostray's election at the meeting of the New England Foundrymen's Association on December 13.

On motion of Past President E. H. Ballard, seconded by Past President N. K. B. Patch, the following Life Members were proposed for election to Honorary Membership at the next Annual Meeting of the Association:

John Howe Hall—Whiting Medalist, 1924.
Enrique Toueeda—Penton Medalist, 1924.
E. V. Ronceray—Whiting Medalist, 1926.
John Shaw—Penton Medalist, 1926.
Harry A. Schwartz—Penton Medalist, 1930.
Ralph S. MacPherran—Whiting Medalist, 1931.
H. W. Gillett—McFadden Medalist, 1932.

A.F.A. Award Donors' Lectures

At the evening session Secretary Hoyt submitted for consideration copy of letter to President Lanahan under date of October 18. In this letter it was pointed out that one of the conditions of the agreement between the donors of Award Funds and the Association, for the use of the income from these funds reads as follows:

"To engage the services of such persons as may be selected by the Board of Awards, to expound technical subjects in which the Association is interested, at its Annual Convention."

Mention was then made of the Marburg Lecture of the A.S.T.M., the

Campbell Memorial Lecture of the A.S.S.T., and the Howe Memorial Lecture of the A.I.M.E., and others of this character sponsored by technical and scientific societies, and the suggestion was made that the Board of Awards be asked to consider the question of providing an annual lecture which might be designated "The A.F.A. Award Donors Lecture."

Following discussion it was agreed to recommend to the Board of Awards their consideration of establishing a memorial lecture which would probably be delivered at the business meeting of the Association.

Respectfully submitted,

C. E. HOYT, *Executive Secretary.*

Auditor's Report

July 18, 1933.

General Thomas S. Hammond, President,
American Foundrymen's Association, Inc.,
Chicago, Ill.

DEAR SIR:

I have examined the books of the American Foundrymen's Association, Inc., for the year ending June 30, 1933, and now submit the following statements:

Exhibit A—Balance Sheet, June 30, 1933.

Exhibit B—Income and Expenses for the year.

Exhibit C—Surplus, June 30, 1933.

Schedule I—Award Funds, June 30, 1933.

Schedule II—Reserve Fund, June 30, 1933.

Schedule III—Cash Receipts and Disbursements for the year.

The securities in the Reserve Fund as shown by the Balance Sheet are stated at cost and no adjustment has been made for shrinkage in their market value.

Amounts receivable include \$2,754.50 for dues of members for the past year and which are of doubtful value.

Subject to these exceptions, it is my opinion that the Balance Sheet at June 30, 1933, shown in Exhibit A attached, correctly reflects the condition of the Association at that date and as shown by the books.

Respectfully submitted,

ROBERT T. PRITCHARD,
Certified Public Accountant.

NOTE: A copy of the Auditor's Balance Sheet and Statement of Cash Receipts and Disbursements is shown in the following pages.

BALANCE SHEET AS AT JUNE 30, 1933

ASSETS

June 30, 1933

ASSOCIATION ASSETS

Furniture and Fixtures, less Depreciation Reserve.....	\$ 387.89
Cash in Bank, Schedule III.....	1,317.04
Accounts Receivable	
Dues	2,754.50
Space Rentals and Permits.....	1,299.50
Other	311.86
Supplies on Hand.....	833.20
Prepaid Expense on Account of Exhibits.....	895.22
Tickets—Century of Progress.....	717.40
	<hr/>
	8,516.61

RESERVE FUND (Schedule II)

Investments (Market Value—\$24,429.00).....	26,749.85
Cash in Bank on Savings Account.....	333.86
	<hr/>
	27,083.71

PHILADELPHIA RESEARCH FUND

Cash in Bank on Savings Account.....	1,103.21
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STEEL CASTINGS TEST FUND

Cash in Bank on Savings Account.....	298.94
	<hr/>
TOTAL.....	\$37,002.47

LIABILITIES

June 30, 1933

ASSOCIATION LIABILITIES

Unpaid Bills	\$ 2,240.82
Dues Paid in Advance.....	350.75
Exhibitor's Permits, etc., Paid in Advance.....	200.00
Unexpended Appropriations for Research	
Sand	486.07
Steel Castings Surfaces.....	1,350.00
Phosphorus and Sulphur in Steel.....	100.00
	<hr/>
	4,727.64
Surplus per Exhibit C.....	3,788.97
	<hr/>
	8,516.61

RESERVE FUND (Schedule II)

Principal of the Fund.....	27,083.71
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PHILADELPHIA RESEARCH FUND

Balance of Subscription	1,103.21
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STEEL CASTINGS TEST FUND

Unexpended Appropriation	298.94
	<hr/>
TOTAL	\$37,002.47

CASH RECEIPTS AND DISBURSEMENTS

For the Year Ending June 30, 1933

CASH IN BANK, July 1, 1932..... \$ 2,671.89

RECEIPTS

Dues and Subscriptions.....	\$21,002.25
Registration Fees	732.75
Space Rentals	7,550.00
Permits	975.00
Award Funds—Balance Due.....	15.00
Appropriated from Reserve Fund	
Principal	2,996.25
Interest	1,953.91
Philadelphia Research Fund	
For Account of Handbook.....	1,001.48
Interest	18.13
Total Receipts.....	<u>36,244.77</u>
	<u>\$38,916.66</u>

DISBURSEMENTS

Salaries	\$19,665.88
Committee Expense	963.77
Handbook	1,001.48
Publications less Sales.....	5,846.48
Printing	933.00
Postage	1,549.52
Office Rent	2,340.00
Telephone and Telegraph.....	390.95
Office Expense	345.14
Exchange and Tax.....	137.86
General Expense	69.58
Travelling Expense	326.76
Insurance	22.51
Dues—Other Associations	80.50
Exhibit Advertising	761.40
Audit	220.00
Convention Expense	1,017.59
Exhibit Expense	995.70
Interest Paid	18.95
Duplicator Purchased	44.10
Century of Progress Tickets Purchased*—	
less sold to date.....	717.40
Sand Research	151.05
Total Disbursements.....	<u>37,594.62</u>
	<u>\$ 1,317.04</u>

CASH IN BANK, June 30, 1933.....

* Redeemed in full after July 1, 1933.

By-Laws

(Approved Feb. 8, 1934)

ARTICLE I

Section 1—This Association shall be known as the American Foundrymen's Association, Incorporated.

Section 2—The objects of this Association as outlined in its articles of incorporation are to promote the arts and sciences applicable to metal casting manufacture and to improve the methods of production and the quality of castings to the end that the increasing utility of all classes of castings may result advantageously to all persons engaged in the foundry and related industries and to all users of foundry products.

The methods for accomplishing the above objectives shall be as determined from time to time by the Board of Directors.

ARTICLE II

Membership

Section 1—The membership of this Association shall consist of General Firm Members, Limited Firm Members, Personal Members, Affiliate Members, and Associate Members, Honorary Life Members, and Life Members.

Section 2—**GENERAL FIRM MEMBERSHIP**—Any firm, organization, or proprietor interested in the objects of this Association shall be eligible for General Firm Membership. A General Firm Member shall be privileged to have one or more subsidiary plants or branches made General Firm Members, also to have one or more associated persons made Affiliate Members, as hereinafter provided.

Section 3—**LIMITED FIRM MEMBERSHIP**—Any firm, organization, or proprietor interested in the objects of this Association shall be eligible for Limited Firm Membership, with membership privileges exercised by a designated individual. The privileges of such membership shall not include for associated persons the right of election to Affiliate Membership.

Section 4—**PERSONAL MEMBERSHIP**—Any person interested in the objects of this Association shall be eligible for Personal Membership. Such membership is non-transferable. No individual by virtue of Personal Membership shall be considered as representing a firm or organization in this Association.

Section 5—**AFFILIATE MEMBERSHIP**—Any person associated with a firm, organization, or proprietor holding General Firm Membership shall be eligible for Affiliate Membership when such person is regularly employed where a General Firm Membership is held.

Section 6—**ASSOCIATE MEMBERSHIP**—Any person interested in the objects of this Association and engaged chiefly in educational or research work, not directly related commercially to casting manufacture, shall be eligible for Associate Membership.

Section 7—**ELECTION TO MEMBERSHIP**—Application for membership shall be made in writing and election to any of the above classes of membership shall be by a three-fourths vote of all Directors, taken by letter ballot.

Section 8—**HONORARY LIFE MEMBERSHIP**—Honorary Life Members shall be persons of acknowledged eminence or who have rendered exceptional service to this Association. Election to Honorary Life Membership shall be by a three-fourths vote of the members of the Association present at a regular meeting thereof, voting on a recommendation of the Directors that the Honorary Life Membership be conferred.

Section 9—**LIFE MEMBERSHIP**—Life members shall be persons who have rendered long or special service to the foundry industry. Election to Life Membership shall be by a three-fourths vote of the Directors.

Section 10—**PRIVILEGES OF MEMBERSHIP**—All members shall have the right to hold office and to attend all regular A.F.A. convention sessions,

and shall receive gratis the regularly issued Transactions of the Association published during each year in magazine or pamphlet form.

Each member in any class shall be entitled to cast one vote on all questions submitted to the membership for letter or other ballot. Each firm or organization member shall designate one individual who shall exercise said privileges for such membership, and who shall be entitled to one vote. The designation of the person so accredited shall be in writing in such form as the Directors shall determine. It shall continue in force until such membership lapses or until authorization is cancelled by the firm or organization.

Section 11—Resignations shall be submitted in writing to the Executive Secretary. The resignation of a member whose dues are in arrears shall not be accepted.

Section 12—The membership of any person, firm or organization may be suspended or terminated by a three-fourths vote of all the Directors. In such case remission of proper proportion of dues paid in advance shall be made.

ARTICLE III

Dues

Section 1—Annual dues for members in the United States and Canada shall be as follows:

General Firm Members.....	\$25.00
Limited Firm Members.....	15.00
Personal Members.....	10.00
Affiliate Members.....	6.00
Associate Members.....	5.00

The annual dues for General Firm Membership shall be \$15.00 when such membership is held by a branch or subsidiary unit of an organization that holds General Firm Membership and pays therefor annual dues of \$25.00.

Section 2—Annual dues for any member regularly established or resident outside of the United States and Canada shall be \$10.00.

Section 3—Honorary Life Members and Life Members shall be exempt from payment of dues.

Section 4—All dues shall be paid to the order of the Treasurer of the Association. The Board of Directors may fix the date on which all dues of members shall become payable and may pro rate or adjust dues of newly elected members.

Section 5—Non-payment of Dues: Any member shall automatically cease to be a member of the Association, whose dues to the Association shall remain unpaid for a period of one calendar month after "final bill" for the same shall have been mailed to the member or to its representative, if the member be a firm or corporation.

ARTICLE IV.

Officers and Directors

Section 1—The officers shall consist of a President and a Vice-President, elected annually from and by the members; and an Executive Secretary, a Treasurer, a Technical Secretary, an Assistant Secretary, and a Manager of Exhibits, all of whom shall be elected annually, when practicable, by the Board of Directors. Any offices, except those of President and Vice-President, in the discretion of the Board of Directors, may be combined.

Section 2—The Board of Directors shall consist of the President, the Vice-President, and fifteen other Directors, elected by and from the members.

Section 3—A President and a Vice-President shall be elected annually to serve for one year, or until their successors are elected and qualified.

Five Directors shall be elected annually to serve for a term of three years, or until their successors are elected and qualified.

Section 4—In the event of a vacancy occurring in the Board of Direc-

tors by death, resignation, promotion by election as President or Vice-President, or for any other reason than retirement at the end of three years' service, the remaining members of the Board shall elect a successor to fill the vacancy and to serve for the unexpired term. A member of the Board whose term has not expired and who is elected President or Vice-President, shall be considered to have vacated the former office held by him, and the Board shall fill the vacancy as above provided.

Section 5—In the event of a vacancy occurring in any office except that of President, the Board of Directors shall fill the vacancy for the unexpired term.

ARTICLE V

Meetings

Section 1—There shall be an annual Convention of this Association, the date and location of which shall be fixed by the Board of Directors at least three months in advance. There shall be an annual business meeting of the Association during the annual Convention. Intermediate meetings of the Association may be held if authorized by a three-fourths vote of all the directors. Twenty-five members shall constitute a quorum of the Association.

ARTICLE VI

Duties of Officers

Section 1—The duties of the President shall be to preside at the meetings of the Association and of the Board of Directors and to perform such other duties as usually devolve upon the chief executive officer. He shall be ex-officio, a member of every committee except the nominating committee.

Section 2—The Vice-President shall perform the duties of the President when the latter is absent or unable to perform the same, and he shall become President in case of a vacancy in that office.

Section 3—The Executive Secretary shall be elected annually by the Board of Directors at the annual meeting of the Board, and shall hold office for one year, or until his successor is elected and qualified.

He shall be, under the direction of the President and Board of Directors, the Executive officer of the Association, the Board of Directors, and its Executive Committee, shall prepare the business therefor, and duly record the proceedings thereof.

He shall see that all moneys due the Association are carefully collected. He shall personally certify the accuracy of all bills and vouchers on which money is to be paid.

He shall present annually to the Board of Directors a balance sheet of his books as of the 30th of June, and shall furnish from time to time such other statements as may be required of him.

He shall perform other duties which may from time to time be assigned to him by the Board of Directors.

Section 4—The Treasurer shall receive all moneys and deposit same in the name of the Association. He shall invest all funds not needed for current disbursements, as shall be ordered by the Board of Directors. He shall pay all bills when certified to by the President and Executive Secretary. He shall make an annual report and such other reports as may be prescribed by the Board of Directors. The Treasurer shall give a bond, the amount of which shall be fixed by the Board of Directors, the premium on said bond to be paid by the Association.

Section 5—The duties of all officers or employees of the Association, not defined in these By-Laws, shall be as prescribed by the Board of Directors.

Section 6—An annual audit of the books and accounts of the Executive Secretary and of the Treasurer shall be made as of date of June 30th, by certified public accountants to be named by the President, and said audit shall be published to the membership.

Section 7—The general control of this Association shall be vested in

the Board of Directors, who shall manage the affairs of the Association in conformity with the laws under which the Association is incorporated, and the provisions of these By-laws.

Section 8—The Board of Directors shall have the power to determine the form and method of distribution of the publications of the Association.

ARTICLE VII

Salaries

Section 1—The Board of Directors shall annually make provision for the salaries of officers, and compensations of all persons whose services to the Association justify such action. Information concerning all salaries and compensations paid shall be published to the membership.

ARTICLE VIII

Nominations of Officers and Directors.

Section 1—There shall be created annually a Nominating Committee consisting of the last three living past presidents of the Association and four other members, elected by the members of the Association at the annual meeting preceding the year of their service.

Section 2—The names of the Nominating Committee shall be announced to the members by letter or in one of the publications of the Association at least four months prior to the annual meeting of that year. If for any reason there shall be a vacancy or vacancies in the Nominating Committee after election, said vacancy or vacancies shall be filled by the President. The senior past president shall be the chairman of the committee.

Section 3—Duties of Nominating Committee: On any day at least fifteen full weeks prior to the annual business meeting, the Nominating Committee shall meet at a time and place designated by the Chairman and shall name one candidate for each office and for each directorship that shall become vacant at the close of the annual meeting of the Board of Directors held in accordance with the provisions of these By-laws.

Section 4—Report of Nominating Committee: Immediately after the candidates are thus nominated, the Chairman shall report the names of the nominees to the Executive Secretary and said report shall be sent by the Executive Secretary to all members of the Association at least ninety days before the annual business meeting.

Section 5—Additional Nominations: After the report of the Nominating Committee has been published and at any time sixty days prior to the date of the annual business meeting additional nominations for any or all vacancies may be made by petition filed with the Executive Secretary and signed by 25 members in good standing.

ARTICLE IX

Elections

Section 1—Should no other candidates for officers and directors be nominated in the manner provided in Article VIII, Section 5, nominations shall be closed and the secretary shall, at the annual business meeting, cast the unanimous ballot of all members for the election of the candidates named in the report of the Nominating Committee.

Section 2—In the event of there being additional candidates nominated for officers and directors in the manner provided in Article VIII, Section 5, then the election of officers and directors shall be by mail ballot of the members.

Section 3—Judges of Election: In the case of election by mail ballot it shall be the duty of the President to appoint three judges of election who shall have sole charge of the election. The judges shall direct the Executive Secretary to have printed and mailed to all members in good standing, ballots bearing the names of all candidates nominated in accordance with provisions of these By-laws. The judges shall name the place of balloting and the final dates for ballots being received.

Section 4—When polls are closed the judges shall examine all ballots cast and shall certify the result of the election to the President before the date set for the annual business meeting of the Association.

The newly elected officers and directors shall assume office at the annual meeting of the Board of Directors following their election as provided for in the order of business in Article XI.

ARTICLE X *Executive Committee*

Section 1—At the annual meeting of the Board of Directors the Board shall elect four of their members who, together with the President, Vice-President and Executive Secretary, shall constitute an Executive Committee, with power to act for the Directors in the interim between meetings of the Board, in all matters which may thus be decided with propriety.

ARTICLE XI *Board Meetings*

Section 1—The annual meeting of the Board of Directors shall be held within ninety days of the date of adjournment of the Annual Business Meeting of the Association, at a time and place designated by the President.

Section 2—The order of business at the annual meeting of the Board of Directors shall be as follows:

- Reading of Minutes.
- Reports of Officers.
- Unfinished Business.
- Recommendations to New Board of Directors.
- Adjournment of Old Board.
- Installation of New Officers and Directors.
- Organization of New Board.
- Election of Officers as Provided for in Article IV, Section 1.
- Election of Four Members of Executive Committee.
- Fixing of Salaries and Compensations.
- Appointment of Standing and Special Committees.
- New Business.
- Adjournment.

Section 3—Special meetings of the Board of Directors may be called at any time by the President, and shall be called by him on the written request of any three members of the said Board. Written notice shall be forwarded to each Director not less than five days prior to any meeting. Seven members of the Board shall constitute a quorum thereof.

ARTICLE XII *Advisory Board*

Section 1—All Past Presidents of this Association shall constitute an Advisory Board, and as members thereof shall have the right to be present and offer advice at all meetings of the Board of Directors. They shall be regularly notified of all Board meetings.

ARTICLE XIII *Board of Awards*

Section 1—The Board of Awards (as established in 1922) shall consist of the last seven living past presidents of the Association, which Board will be self perpetuating as each retiring president becomes a member taking the place of the senior past president on said Board. The junior past president shall be the chairman.

Section 2—The Board of Awards shall control the disposition and investment of Award funds, together with the income therefrom, subject to the terms and conditions set forth in agreement with the donors of special award funds.

ARTICLE XIV

Prizes

Section 1—The Directors may from time to time under regulations which they may adopt offer medals or prizes for exceptionally meritorious investigations or developments which benefit the foundry industry. The cost of these medals or prizes may be defrayed from funds of the Association or from any funds which may be given to the Association for such purpose.

ARTICLE XV

Divisions

Section 1—Groups representing the various branches of the casting industry, to be known as Divisions of the Association and organized from its members may be authorized by the Board of Directors, and any member of the Association may register for membership in any of the Divisions in which he is interested. A manual for the conduct of Divisions, approved by the Directors, shall govern the activities and procedure of all Divisions.

ARTICLE XVI

District Organizations

Section 1—For the purpose of promoting the objects of the Association the Directors may establish Districts that can be satisfactorily served through occasional meetings of members residing therein, and through joint meetings with other technical or industrial organizations.

The Directors shall determine the plan of district organization, define the territory of each district, and shall have the authority for making all arrangements for holding and financing district meetings.

ARTICLE XVII

Local Sections.

Section 1—Local sections of the Association may be authorized by the Directors' acceptance of the written request of a satisfactory number of members residing within a prescribed territory who desire to hold regular meetings for promoting the objects of the Association. Only one Local Section shall be authorized in any locality. Such Local Section shall be designated as the Section of the American Foundrymen's Association, Inc.

Section 2—The provisions of the By-laws of the Association and of a manual for conducting Local Sections, approved by the Directors, shall govern the procedure of all Local Sections. No action or obligation of a Local Section shall be considered an action or obligation of this Association. The Board of Directors shall have the right at any time to dissolve a Local Section for good and sufficient reasons, or to readjust its boundaries after serving sixty days' notice of its intention to do so.

Section 3—All members of the Association residing within the prescribed territory of any Local Section shall be eligible for membership in such section. They shall not, however, be enrolled as members of the Local Section unless they have signified in writing their desire to be so enrolled.

Section 4—For financing Local Sections the Board of Directors *may* contribute from its funds an annual stand-by charge for each Local Section that shall not exceed \$50.00 each year, and may authorize an additional appropriation which shall not exceed each year 15 per cent of the annual dues received from registered resident members of said Local Section in said year. If the expenses of a Local Section exceed the appropriations made by the Board, the difference may be made up by voluntary contributions or in other ways by the members of the Local Section as they may determine. The Association shall not be responsible for the debts of any Local Section.

ARTICLE XVIII.

Seal and Emblem.

Section 1—The emblem of the Association shall be of circular form with the initials and date of organization of the Association occupying the cardinal points and connected by double rings. Crossing said rings at right-angles, and joined in the center by a vertically placed trowel, a hammer, and a shovel shall be represented.

Section 2—Pins or other articles bearing the emblem of the Association may, in the discretion of the Board of Directors, be issued or sold to members.

Section 3—The seal of the Association shall be a reproduction of the emblem, with sufficient additional space on the circumference to show these words surrounding the emblem proper: "Seal of American Foundrymen's Association, Inc."

ARTICLE XIX.

Standards

Section 1—No detail of foundry practice or design of foundry appliance shall be adopted as a standard of the Association or as recommended by it except subject to the approval of a three-fourths vote of the Directors. No appliance or item of equipment shall be adopted as a standard of the Association or recommended by the Association to the exclusion of similar equipment if it is manufactured or sold only by one firm or commercially restricted group of firms.

ARTICLE XX.

Amendments

Section 1—These By-laws may be amended only by a majority of votes cast by letter ballots which shall be submitted by the Executive Secretary to all members in good standing. Said letter ballots must bear signatures of all those voting and they must be received at the Executive Secretary's office within thirty days after being mailed therefrom to the members, in order to be included in the canvass of votes. Preliminary to the submission to the membership of said letter ballots, action by a two-thirds vote of the members present at a business meeting of the Association or action by a two-thirds vote of the Board of Directors at any meeting thereof shall take place, favoring any amendment to be submitted. The Executive Secretary shall mail letter ballots to all members in good standing within thirty days after adjournment of the meeting authorizing their submission.

ARTICLE XXI

Rules of Order

Section 1—Roberts' Parliamentary Rules of Order shall be recognized as authority by this Association, and shall govern the deliberations in all cases not covered by these By-laws.

**Principles of Practices of the American Foundrymen's Association**

(Approved by Board of Directors, Dec. 9, 1933)

The fundamental purposes of the American Foundrymen's Association, as set forth at the birth of the organization in 1896 and restated under the objects listed in the charter of incorporation granted by the State of Illinois in 1916, fall short in defining the scope of recent activities and the field of service demanded from the organization by modern conditions.

Accordingly, new principles of practice to reaffirm the responsibilities and to set forth the duties of the American Foundrymen's Association are offered by the Board of Directors on behalf of the membership, as follows:

The American Foundrymen's Association is a voluntary organization of individuals and firms in the foundry and allied industries, joined together to promote and improve the production and application of all cast metals and alloys.

The improvement in production methods results from advancement in technical practices through the association's efforts in providing stimulating papers, discussions and instruction in the sciences, arts and crafts of castings manufacture. The association encourages and sponsors active research to improve materials, processes and methods. It stimulates exchange of knowledge and centers attention on new developments through holding periodic meetings comprising technical sessions, round table conferences and shop instruction courses. Intimate acquaintance with new materials and equipment is offered through exhibitions, displays and demonstrations arranged concurrently with these conventions. It prints, publishes and distributes formal papers and resulting discussions relating to technical and association affairs. In all activities relating to advancement in various phases of the foundry industry the association acts without favor or prejudice as an impartial agency for the dissemination of information.

To promote the wider use and service of castings, the American Foundrymen's Association extends its full facilities as an agent by and for the foundry industry. Co-operating with recognized and approved organizations, it always is diligent in advancing specifications in co-operation with producers and consumers of cast products and their component materials. Its precise and far reaching knowledge of the engineering utility of cast products always is available to the engineering profession individually or through groups and associations. The encouragement of symposia on the four main branches of cast metals is a fine example of this responsibility fulfilled by the A.F.A. Further, the association is a fair and just champion of cast products as opposed to unwarranted commercial claims of competing commodities outside the foundry industry.

As a further aid to the advancement of castings use, the American Foundrymen's Association has co-operated in the past and will continue to co-operate with trade and business groups of foundrymen in all matters wherein their interests do not conflict with each other. The organization of two of the major trade associations of foundrymen through meetings called coincident with annual conventions of the American Foundrymen's Association, the active participation in aiding these and other such organizations in attaining their laudable aims for the benefit of their individual branches of the industry constitute examples of this phase of A.F.A. service. Under the new integration of all industry, the American Foundrymen's Association faces new tasks in relation to its prime purpose as a technical organization. The association reaffirms its willingness and ability to foster and advance all matters relating to the science and practice of castings manufacture for each of the several branches of the industry.

The Action of Coal Dust as a Facing Material

By BEN HIRD,* NEWPORT, MON., ENGLAND

Abstract

To obtain good results from coal dust additions to molding sands, right quality material is essential. For all classes of work, the coal should be of a highly volatile nature and of fine grain. The volatile material should not be less than 28 per cent. Correct proportions of coal well mixed with the sand will improve the skin of the casting and promote clean stripping by imposing a carbon film between the molten metal and the mold face. It will assist in preventing "sand scabs" and "drawing down" by releasing liquefied tars which bind the sand grains together as the mold fills. It also produces a more refractory sand by coating the grains with a carbon deposit. Further, coal creates a green bond which is developed by milling and which appears to be due to rough carbon deposit on the sand grains and a tarry substance distilled from the coal.

INTRODUCTION

1. It is not the object of this paper to lay down any hard and fast laws relating to the action of coal dust on molding sands, but rather to put forward the author's conclusions formed during a period of investigation into the subject of coal dust, carried out in an ordinary gray iron foundry without laboratory equipment. The object is to arouse interest in this worthy but rather neglected subject of coal dust additions to molding sands by describing some experiments along these lines.

QUALITIES OF COAL DUST

2. The coal used should be high in volatile matter because it is the dominant element; from its virtue the value of coal dust additions to sand is derived. Anthracite coal is useless as a facing. Double the quantity of such low-volatile coal is required to

* Foundry Manager, Cwmbran Works, Keen & Nettlefolds, Ltd.

NOTE: This paper was presented and discussed before one of the sand sessions at the 1933 Convention of the American Foundrymen's Association.

get even approximately the sand results as with the high-volatile material; *i.e.*, using 10 per cent of a 15-per cent volatile coal will not product as good results as using 5 per cent of a 30-per cent volatile coal.

3. The fineness of coal dust is equally important. As a minimum requirement, it should all pass through a sieve with 40 meshes to the linear inch. The finer it is, the better it will be distributed among the sand grains, and the more easily will it be volatilized by the heat of the metal. Using fine coal, the skin of the casting will be smoother and freer from pockmarks caused by the swelling and coking of large specks of coal, especially in light castings, than if coarse coal is used.

MIXING

4. The mixing should be thorough for an even distribution of the coal dust in the sand. Theoretically, each grain of sand should have specks of coal dust adhering to it or near to it. Patches of undistributed coal will cause mapping and other troubles.

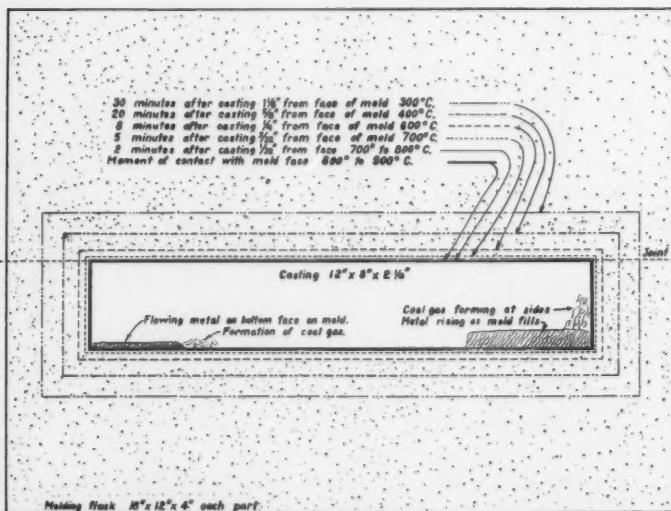


FIG. 1—MOLD DIAGRAM (MASKE AND PIOWOWARSKY), SHOWING EFFECT OF CASTING HEAT IN RAISING TEMPERATURE OF ADJACENT SAND AT VARIOUS TIME INTERVALS AFTER COOLING.

ACTION OF MOLTEN METAL ON COAL DUST AND SAND IN MOLD

5. Taking the test block shown in Fig. 1 as an example, the following action takes place when the mold is poured. Assuming the mold has been made of well-mixed sand, with the proper percentage of good coal dust, instantly on the approach of the metal these specks of coal give off their volatiles in the form of gas or smoke. This is trapped between the rolling flow of the metal and the mold face, thus protecting it by a thin film of carbon.

6. This action takes place rapidly on the bottom face. On the sides, the specks of coal volatize over a larger area in the front of the rising metal because of the greater radiation of heat, and the carbon from the escaping gases is deposited in the form of soot on the upper portions of the sides and top of the mold, there forming an extra protection to the sand.

7. This sudden intense heat which gasifies the face coal, liquefies the volatiles in the coal immediately behind the face into the form of tar, which liquefying material, due to the influence of escaping gases, swells and spreads over the grains of sand, bonding them together. This concreting action adds considerable reinforcement to the sand forming the face of the mold, which is further strengthened when the liquid tars are carbonized or coked by the penetration of higher temperatures.

8. Thus, this action of the coal dust creates a "hot bond" over the face of the mold which resists "scabbing" and "drawing down." Immediately behind these face grains, the sand and coal dust is as yet unchanged and provides escape for the gases that pour through it. To these gases the sand acts as a scrubber or filter, condensing and collecting the tar and carbon, with which they become coated. This action continues out into the sand until the limit of heat penetration is reached. Reference to the heat penetration diagram, Fig. 1, made from results given by F. Maske and E. Piwowarsky, may assist in following these statements.

9. Increased refractoriness is demonstrated by experiments, and it is fairly obvious that sand grains coated with carbon are more refractory than clear sand grains or those coated with clay bond. If sufficient coal dust is added to the sand after each cast to counteract the burning-off of this coating, a sand can be maintained that will give a good skin to the castings and satisfactory stripping qualities. Furthermore, it will "create green bond," especially if the sand is milled.

EXPERIMENTAL WORK

10. The author has carried out a number of experiments proving this statement regarding the creation of green bond, and also has put the results in practical application in the foundry, which has produced a considerable saving of new sand. He cannot give any definite theory as to how this bond is created, but tentatively advances the following data.

11. The sand grains become coated with a rough carbon deposit which, combined with a small percentage of distilled tar from the coal dust, creates a bond, partly frictional (*i.e.*, the gripping of the rough surfaces) and partly glutinous (*i.e.*, tarry residue). In his opinion, this bond will provide the base for synthetic sand, which can be further strengthened if required by the addition of very small quantities of special bonding material.

12. Table 1 gives the details of material and test apparatus used in our experimental work.

Experiment 1—Natural Bonded New Molding Sand.

13. In this first test, Stourbridge red sand was used with 10 per cent coal dust. The mixture had 5 per cent moisture. Tests on both milled and hand-mixed sands were made. In the milled

Table 1
COAL DUST BOND EXPERIMENTS

I—Materials Used

- (a) Molding Sand: Stourbridge red sand (as quarried).
- (b) Sharp Sand: Briton Ferry dune sand.
- (c) Burnt Sand: Old black sand from Stourbridge red. Carbon coating burnt off.
- (d) Coal Dust: Volatile 30.6, Ash 10.4, Carbon by diff. 50 per cent. Fineness, all through 90-mesh.

II—Test Specimen

- (a) Test Casting: Rectangular block 12 x 8 x 2½ in., weighing 64½ lbs. Made in molding flask, pressed steel, no bar, 16 x 12 x 4 in. each part.

III—Sand Testing Apparatus

- (a) Richardson's permeability.
- (b) A.F.A. drop ram and spring balance compression. 2 x 2 in. test piece.
- (c) Check results taken on A.F.A. permeability apparatus design and compression strength of Dietert design.

Table 2

**EFFECT OF COAL ON STOURBRIDGE RED SAND—HAND MIXED
(10% COAL DUST ADDED TO ORIGINAL MIX,
NO FURTHER ADDITIONS MADE)**

Original Sand:		Coal Dust.....	10%				
Test No.	Per Cent Coal Dust Added.	Richardson Permeability.....	3½				
		Richardson Strength.....	13				
Per Cent — Tests Using Richardson Apparatus — Mold Condition*, in. — Condition of Casting.							
		Permeability	Strength	"A"	Silvery	Black	
1	Nil	2½	8	½	2½	1½	Casting good. Good blue skin.
2	Nil	2½	8	Nil	2½	1½	Casting good. Gray blue skin.
3	Nil	2½	7½	Nil	1½	1½	Casting good. Rough gray skin.
4	Nil	2½	7½	Nil	1½	1½	Casting good. Rough gray skin.
5	Nil	2½	6	Nil	1	1	Casting good. Rough gray skin.
6	Nil	2½	5	Nil	1	1	{ Top face drawn down slightly. Very rough gray skin.

*Condition of sand adjacent to casting. See Fig. 7.

Table 3

**EFFECT OF COAL ON STOURBRIDGE RED SAND—MILLED
(10% COAL DUST ADDED TO ORIGINAL MIX,
NO FURTHER ADDITIONS MADE)**

Original Sand:		Per Cent Coal Dust Added.	Richardson Permeability.....	Strength	Mold Condition*, in. —	Condition of Casting.
Test No.	Treatment.	After Milling }	10	3½	23	
1	{ Mixed	Nil	2½	14	{ Nil	{ Top face large blind scale. Sides good. Bottom bad wave mark. Blue skin.
	{ Milled	Nil	3½	20	{ Nil	{ Good casting. Gray blue skin.
2	{ Mixed	Nil	3½	15	{ Nil	{ Good casting. Gray skin.
	{ Milled	Nil	3½	20	{ Nil	{ Good casting. Gray skin.
3	{ Mixed	Nil	3½	14	{ Nil	{ Good casting. Gray skin.
	{ Milled	Nil	3½	21	{ Nil	{ Good casting. Gray skin.
4	{ Mixed	Nil	2½	11	{ Nil	{ Good casting. Gray blue skin.
	{ Milled	Nil	3½	21½	{ Nil	{ Good casting. Gray blue skin.
5	{ Mixed	Nil	3½	14	{ Nil	{ Good casting. Gray blue skin.
	{ Milled	Nil	3½	23	{ Nil	{ Good casting. Gray blue skin.
6	{ Mixed	Nil	3½	14½	{ Nil	{ Good casting. Gray blue skin.
	{ Milled	Nil	3½	22½	{ Nil	{ Good casting. Gray blue skin.

*Condition of sand adjacent to casting. See Fig. 7.

sand series the mixture was milled 3 minutes before each casting was made. In the hand-mixed series the mixture was well turned and put through a $\frac{1}{8}$ -in. sieve twice. The tabulated results of this test are given in Tables 2 and 3, which show permeability and strength for the various series.

14. The results of the hand-mixed series (Table 2) appear to be about normal. The addition of fine coal dust decreases permeability and increases strength slightly, which then decreases after each cast, while the permeability increases.

15. In striking contrast, the milled series (Table 3) shows an increase in strength after each cast when re-milled. Reviewing this feature, it appears obvious that the sand is gaining bond from some source other than the clay bond or alumina of the red sand grains. Microscopic examination of the sand after the fourth cast showed all the grains covered with a black carbon deposit, which must obviously destroy the clay bond. Yet Test 5 (Table 2) gave a strength equal to the milled red sand. This suggested a bond created by coal dust, and four experiments were made with burned sand.

Table 4
EFFECT OF COAL DUST ON BURNT SAND—HAND MIXED
(10% COAL DUST ADDED TO ORIGINAL MIX,
NO FURTHER ADDITIONS MADE)

Test No.	Per Cent Coal Dust Added.	Tests Using Richardson Apparatus		Mold Condition*, in.		Condition of Casting.
		Permeability	Strength	Silvery "A"	Black "B"	
Burnt Sand Only	Nil	1 $\frac{1}{4}$	2	Nil	Nil	
Mixture Before Use	10	2 $\frac{1}{2}$	4			Top face all pulled down. Bottom swilled Badly. Very bad casting. Extremely rough.
1	Nil	1 $\frac{7}{8}$	5	$\frac{5}{8}$	2	Top face slightly mapped. Bottom badly mapped, swill from pitch of runner. Rough gray skin.
2	Nil	1 $\frac{1}{8}$	4	Nil	1	Good casting. Rough gray skin.
3	Nil	1 $\frac{1}{8}$	4	Nil	1	Good casting. Rough gray skin.
4	Nil	1 $\frac{1}{8}$	4	Burnt clear $\frac{1}{8}$	$\frac{5}{8}$	Good casting. Very rough gray skin. Sand burnt on.

*Condition of sand adjacent to casting. See Fig. 7.

Table 5

EFFECT OF COAL DUST ON SHARP SAND—MILLED
(10% COAL DUST ADDED TO ORIGINAL MIX,
2½% COAL DUST ADDED AFTER EACH CAST)

Test No.	Treatment	Per Cent Coal Dust Added	Tests Using Richardson Apparatus		Check Tests Using A.F.A. Apparatus		Mold Condition*, in.	Condition of Casting
			Permeability	Permanent Strength	Permeability	Compression Strength		
Original Sand	Before Use		10	½	2½	85	1	
1	{ Mixed Milled	2½ 2½	½ 1	2½ 6	85 52	1 2.2	½	1½ Top face badly pulled down. Bottom slightly matted. Rather rough blue skin.
2	{ Mixed Milled	2½ 2½	1 1½	3½ 8½	65 70	1.5 2.2	Nil	1½ Top very slightly pulled down at one corner. Bottom good. Better skin.
3	{ Mixed Milled	2½ 2½	1½ 2½	7 10½	53 49	2.6 3.1	Nil	1½ Good casting. Fair blue skin.
4	{ Mixed Milled	2½ 2½	2½ 2½	12½ 13½	45 40	3.1 4.6	Nil	1 Good casting. Fair blue skin.
5	{ Mixed Milled	2½ 2½	2½ 2½	12 15	30 34	3.8 4.4	Nil	½ Good casting. Fair blue skin.
6	{ Mixed Milled	2½ 2½	3 2½	15½ 16½	32 27	4.8 6.0	Nil	½ Good casting. Fair blue skin.

*Condition of sand adjacent to casting. See Fig. 7.



FIG. 2—SPECIMEN CASTINGS OF EXPERIMENT 3, MADE IN SHARP SAND. 10 PER CENT COAL DUST ADDED TO ORIGINAL MIXTURE. SEE TABLE 5.

Table 6
EFFECT OF COAL DUST ON SHARP SAND—MILLED
(15% COAL DUST ADDED TO ORIGINAL MIXTURE,
5% AFTER EACH SUCCEEDING CAST)

Test No.	Treatment	Per Cent Coal Dust Added	Tests Using Richardson Apparatus		Check Tests Using A.F.A. Apparatus			Mold Condition*, in "A"	Mold Condition of Casting
			Permeability	Strength	Permeability	Compression Strength	Silvery "B"		
Original Mixture}		15	3/4	6	75	1.5			
1	Milled	15	3/4	8 1/2	65	1.7	3/4	3/4	Top and sides good. Bottom slight mapping.
2	{ Mixed Milled	5 5	3/4 3/4	7 9 1/2	70 65	1.5 2.2	{ 1/2	1	{ Casting good. Bottom very slight mapping.
3	{ Mixed Milled	5 5	3/4 1 1/2	7 10 1/2	60 45	1.5 2.8	{ 1/2	1	Good.
4	{ Mixed Milled	5 5	1 1 1/2	7 1/2 12	80 55	1.2 2.4	{ 0	1 1/2	Good.
5	{ Mixed Milled	5 5	1 1/4 2	9 1/2 16	56 45	2.1 3.0	{ 0	1 1/2	Good.
6	{ Mixed Milled	5 5	1 1/2 2 1/2	12 16 1/2	55 40	2.2 4.3	{ 0	1 1/2	Good.

*Condition of sand adjacent to casting. See Fig. 7.

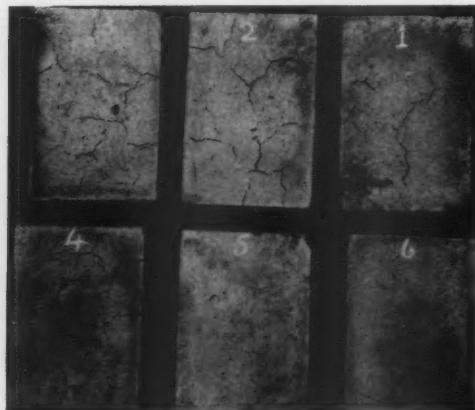


FIG. 3—SPECIMEN CASTINGS OF TABLE 6. CASTINGS ARE SHOWN AS REMOVED FROM THE MOLDS AND CLEANED BY MEANS OF WIRE BRUSHING.

Experiment 2—Use of Burned Sand, Hand Mixed.

16. The same procedure was followed with hand-mixed burned sand, and the results are shown in Table 4. The improvement in the molding properties of the sand and each subsequent casting was very encouraging and better than the test results show. It was decided to make a series of tests using sharp sand with an addition of coal dust after each cast.

Experiment 3—Sharp Sand Series.

17. The material used was sharp sand plus 10 per cent coal dust with a 2½ per cent addition after each cast. It was milled

Table 7
EFFECT OF COAL DUST ON SHARP SAND—HAND MIXED
(15% OF COAL DUST ADDED TO ORIGINAL MIXTURE;
DECREASING AMOUNTS ADDED AFTER EACH SUCCESSIVE CAST,
RANGING FROM 5 TO 1%)

Test No.	Per Cent Coal Dust Added	Tests Using Richardson Apparatus			Tests Using A.F.A. Apparatus			Mold Condition*, in.— "A" "B" Condition of Casting.
		Permeability.	Strength.	Permeability.	Strength.	Compressive Strength.		
Original Sharp Sand	0	½	3½					
Original Mixture	15	1	6½					
1	15	½	6½	60	1.5	½	1	Top face good. Side near runner, slight swill. Bottom mapped, slight swill from pitch of runner.
2	5	½	6½	90	1.0	½	½	Top face good. Sides good. Bottom mapped, swill as above.
3	4	1	7½	60	1.7	½	½	Top face good. Sides good. Bottom mapped, swill. Not so good as 1 and 2.
4	3	½	8½	70	1.2	½	¾	Top face good. Side near runner, slight swill. Bottom, slight swill under runner.
5	2	1	6½	66	1.1	0	1½	Top face good. Sides good. Bottom good. Better casting.
6	1	½	8½	80	Too weak	0	1½	Top face good. Sides good. Bottom good. Skin rather rough. Sand burning on.

*Condition of sand adjacent to casting. See Fig. 7.

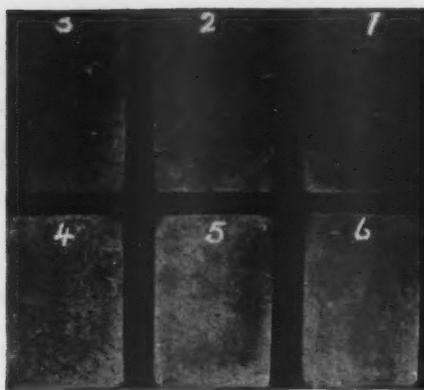


FIG. 4—SPECIMEN CASTINGS OF TABLE 7. REMOVED FROM MOLDS AND WIRE-BRUSH CLEANED.

for 3 minutes with 5 per cent moisture. Table 5 gives permeability and strength results of this series. These were very surprising, but conformed with the improvement in the molding properties of the sand, and resultant castings are shown in Fig. 2.

18. Casting No. 1 of Fig. 2 shows the top face was badly pulled down. The bottom was good but very slightly mapped, while the skin was blue but rather rough. The top of casting No. 2 was very slightly pulled down at one corner. There was a distinct improvement in the skin. Castings Nos. 3, 4, 5 and 6 were all good castings free from defects and having a fairly blue skin.

Experiment 4—Sharp Sand, Milled and Hand Mixed.

19. Another series of tests was made with sharp sand, milled and hand mixed. In the milled series the sharp sand was mixed first with 15 per cent coal dust and 5 per cent coal dust was added after each cast. Great care was taken with this series to insure that no particle of other sand became mixed with the test sand. The time of milling was checked with a stop watch. Three minutes from starting the mill to opening the discharge doors was the standard. It took about one-half minute to discharge the batch.

20. The permeability and strength results are shown in Table 6 and resemble very closely the results of the previous tests. The castings were better and are shown in Fig. 3. In casting No. 1 there was no pulling down of the top face, and except for slight

Table 8
EFFECT OF COAL DUST ON BURNT SAND—HAND MIXED
(15% COAL DUST ADDED TO ORIGINAL MIXTURE,
2½% AFTER EACH SUCCEEDING CAST)

Test No.	Per Cent Coal Dust Added.	Tests Using Richardson Apparatus		Tests Using A.F.A. Apparatus		Mold Condition*, in. "A"	Mold Condition "B"	Condition of Casting.
		Permeability.	Strength.	Permeability.	Com- pression Strength.			
Before Use		15	1½	7½	35	1.7		
1		1	6½	50	1.5	½	1	Top face good. Sides good. Bottom mapped. Swill from pitch of runner.
2	2½	1	6½	40	1.3	½	1½	Top face and sides good. Bottom slightly mapped. Gray blue skin.
3	2½	1	7½	38	1.9	0	1	Casting good. Gray skin. Sand burning on.
4	2½	1	7½	40	2.0	0	¾	Casting good. Gray skin. Sand burning on.
5	2½	1½	8½	37	2.2	0	1½	Casting good. Gray skin. Sand burning on.
6	2½	1½	7½	42	2.0	0	¾	Casting good. Gray skin. Sand burning on.

*Condition of sand adjacent to casting. See Fig. 7.

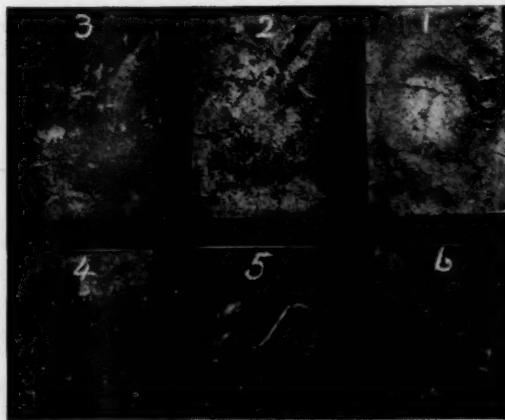


FIG. 5—SPECIMEN CASTINGS OF TABLE 8. CLEANED BY WIRE BRUSH ONLY.

Table 9
EFFECT OF COAL DUST ON BURNT SAND—MILLED
(15% COAL DUST ADDED TO ORIGINAL MIXTURE,
2½% AFTER EACH SUCCEEDING CAST)

Test No.	Per Cent Coal Dust Added	Tests Using Richardson Apparatus		Tests Using A.F.A. Apparatus			Mold Condition* in.— "A" Silvery "B" Black	Condition of Casting
		Permeability	Strength	Permeability	Compression Strength			
Before Use }	15	1½	8½	35	2.2			
1		1½	8½	25	2.4	½	1½	{ Top face slightly mapped. Sides slightly mapped. No swirl. Good skin.
2	2½	2½	11	22	2.7	¼	1	{ Top and sides good. Bottom slight mapping. Good skin.
3	2½	2½	11	35	2.6	0	1½	{ Casting good. Good skin.
4	2½	2½	12	20	2.9	0	1½	{ Casting good. Good skin.
5	2½	2½	13	25	2.8	0	1½	{ Casting good. Good skin.
6	2½	2½	13½	20	2.9	0	1½	{ Casting good. Good skin.

*Condition of sand adjacent to casting. See Fig. 7.

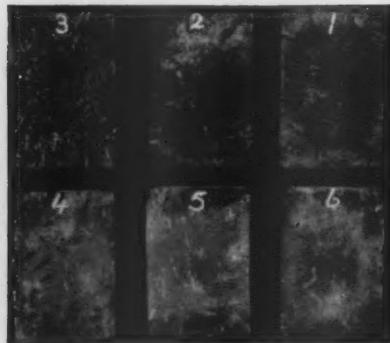


FIG. 6—SPECIMEN CASTINGS OF TABLE 9. CLEANED BY WIRE BRUSH ONLY.

Table 10
FINENESS TEST

SHARP SAND SERIES										BURNT SAND SERIES						STOURBRIDGE RED SAND		
Hand Mixed					Milled					Hand Mixed			Milled			Hand Mixed		
Sharp Sand Only.	No. 1 +5% Coal Dust.	No. 6 +1% Coal Dust.	No. 3 Before Use Coal Dust.	No. 6 +5% Coal Dust.	No. 3 Before Use Coal Dust.	No. 6 +5% Coal Dust.	No. 3 Before Use Coal Dust.	No. 6 +5% Coal Dust.	Burnt Sand Only.	Burnt Sand Only.	Burnt Sand Only.							
Set on 30	0.40	0.31	1.18	0.38	0.42	0.74	0.34	0.56	1.76	0.50	0.90	0.45	0.3	0.3	0.3	0.3	0.3	0.3
Set on 60	18.61	55.1	38.61	30.16	31.30	37.36	8.52	5.63	11.84	7.12	6.82	7.52	13.8	21.0				
Set on 90	66.23	34.06	46.67	57.45	49.66	38.62	37.45	35.83	39.72	58.83	36.76	36.68	34.1	37.0				
Set on 120	12.24	5.97	7.85	7.92	9.66	9.77	26.34	35.94	25.87	8.87	23.80	25.02	27.5	23.5				
Set on 150	0.87	0.72	2.57	0.59	1.21	1.47	11.40	3.30	5.64	9.98	11.83	10.52						
Set on 180	0.53	0.54	0.90	0.56	1.58	1.84	6.70	3.97	4.94	3.41	6.30	5.13	17.2	13.3				
Set on 210	0.13	4.99	5.97	2.69	5.17	4.05	10.45	12.54	12.33	12.48	15.45	14.92	6.1	4.2				

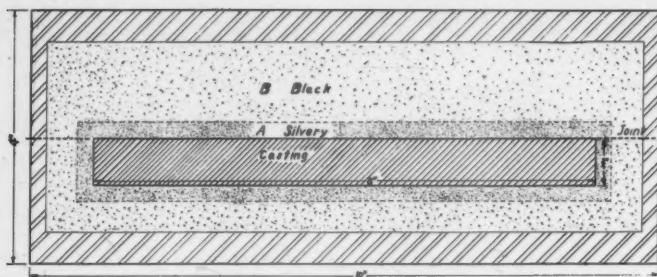


FIG. 7—DIAGRAM OF SPECIMEN MOLD AND CASTING, SHOWING MOLD CONDITION. EFFECT OF HEAT PENETRATION ON COAL FACING IS INDICATED BY THE REGION A WHERE THE SAND HAS A SILVERY APPEARANCE. AMOUNT OF SILVERY SAND IS SHOWN IN VARIOUS TABLES UNDER MOLD CONDITION.

mapping on the bottom face due to excess of coal dust, the casting was quite good. Nos. 2, 3 and 4 were all good castings, mapping decreasing as coal dust burned out, with only a trace in No. 4. Nos. 5 and 6 were perfect castings with a good blue skin.

21. In the hand-mixed material of this series, a gradual reduction was made in the amount of coal dust added after each cast. The results are given in Table 7 and the castings are shown in Fig. 4. The most noticeable features are a gradual decline in the color and smoothness of the casting skin, and a falling off of the "bond strength" in the tests. Again the molding properties of the sand were better than the test results indicate, as instanced by "condition of casting."

Experiment 5—Burned Sand, Milled and Hand Mixed.

22. A further series of tests was made with burned sand, with results as shown in Tables 8 and 9. These included both milled and hand-mixed sands. With the milled sands there was a gradual increase in test strength (Table 9), although not so pronounced as with sharp sand. The castings were quite good (Fig. 6), with the exception of slight mapping caused by excess of coal dust.

23. The tests results of the hand-mixed sands were slightly better (Table 8 and Fig. 5) than the sharp sand, hand mixed, due probably to the addition of consistent amounts of coal dust, although the condition of the castings show there was not sufficient coal dust to maintain a good skin without milling. Judging from results, milling appears to assist the coal dust to function.

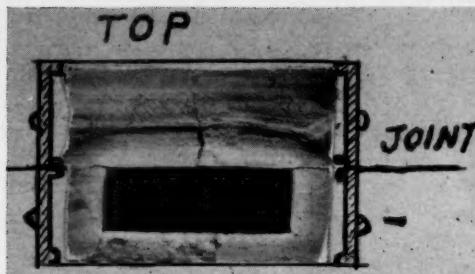


FIG. 8—ACTUAL SECTION OF MOLD AFTER CASTING WAS REMOVED. THIS MOLD WAS ONE OF SERIES OF TABLE 7. LIGHT SECTION NEXT TO THE CASTING CAVITY SHOWS THE SILVERY PORTION OF SAND A OF FIG. 7.

EFFECTS OF FINENESS OF GRAIN

24. Another point of comparison between the burned sand and the sharp sand tests is shown in the sieve test results of Table 10. The fineness of the sand grains and fines from the coke dust were held as a possible explanation of the increase of bond due to coal dust additions, but these results do not uphold this theory. The coarser sharp sand gives the highest bond strength when milled.

INFLUENCE OF HEAT

25. Fig. 7 is a diagram showing the influence of the heat from the casting on the coal dust and sand, while Fig. 8 is a photograph of section through a mold after the casting has been removed. The depth of silvery coking and the tarry bonded black varies with the amount of unchanged coal dust in the sand. These variations are given in the test result tables under the column headed "Mold Condition."

26. The silvery portion "A" is more pronounced in the hand-mixed sands than the milled, and in the sharp sand as compared with the Stourbridge red. It is remarkable that this Stourbridge sand milled showed no sign of silvery coking in Test No. 1, possibly due to the amalgamation of the fine coal with the clay bond.

27. It is interesting to compare this diagram with the temperature diagram of Fig. 1. This is made up from results obtained by F. Maske and E. Piwowarsky.¹ The size of their test casting was 10½ in. diameter and 1¼ in. thick. Therefore, it is probable

¹ F. Maske and E. Piwowarsky, "The Gas Permeability of Molding Sand," *Foundry Trade Journal*, March 28, 1929.

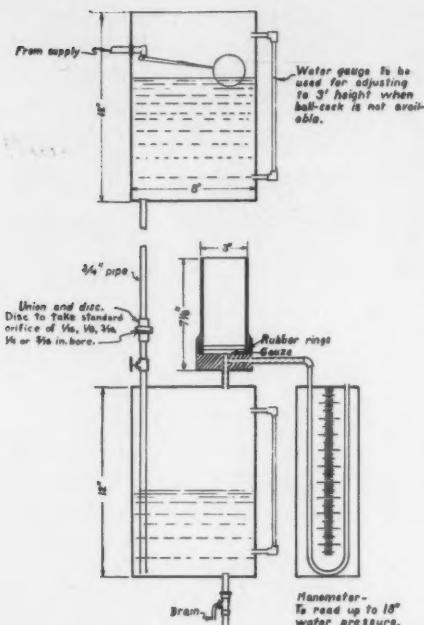


FIG. 9.—DIAGRAM OF RICHARDSON PERMEABILITY TEST APPARATUS.

that the true heat penetration on this diagram would be slightly deeper than the figures show. If full size reproductions of these two diagrams are made on tracing paper and then laid one on the other with the lines of the casting coinciding, an interesting chart is obtained. This was the object in making the diagram to the size of the rectangular test block.

PERMEABILITY APPARATUS

28. In our tests we used Richardson's permeability apparatus (Fig. 9). This consists of two containers, the top acting as a reservoir and the bottom as a compression chamber for the air which is forced through the sample. The top container can be fitted with a ball-cock or gage glass. A water pipe is taken from the base of this container through a stop cock into the base of the bottom container, which is fitted with a drain cock and gage glass. Fixed to the top of this tank is a threaded ring into which the cylinder containing the test piece can be screwed. Two pipes, one

leading into the top of the tank and the other to the water gage, are taken from the bottom of the threaded ring, which has a solid bottom. The sand sample rammed in the cylinder acts as a cork; thus the closer (or less permeable) the sand, the higher the pressure in the tank, which is recorded on a water gage measured off in inches. The $\frac{1}{8}$ -in. orifice was used in our tests.

29. The 2 x 2 in. test piece was given three blows on a standard A.F.A. drop ram machine and the strength taken on a spring balance scale, the pressure being applied with a screw and the results given in pounds as recorded on the standard scale.

DISCUSSION

JOHN GRENNAN: * In presenting this paper for the author, whom we regret could not be present, I wish to say that this paper deals with sea coal from two angles. The first idea of sea coal is that it aids in causing the sand to peel from the casting. This is the most common thought as to the use of this material.

The author first deals with this aspect, showing that the sea coal is effective when it contains high-volatile matter, as the volatile matter is the active agent. The heat drives off the volatile matter in the form of a smoke or gas. It acts as an envelope, or a protective parting, between the metal itself and the sand grains. The smoke also is deposited as a soot on the surface of the mold as it advances ahead of the filling of the mold by the metal.

The author's second point, with which the bulk of the paper deals, is the bonding property of the sea coal. The bonding property of sea coal depends on two factors, (a) the coking of the coal on the sand grains, which coking gives a frictional effect as far as bond is concerned, and (b) the distillation of tars which are deposited on the sand grains.

The author deals with three types of sand. He mixed varying amounts of sea coal with an ordinary molding sand, similar to what we would call a No. 3 in the United States. He next took a sharp, or sand-dune sand, similar to Michigan City sand, with which he mixed varying amounts of sea coal. He finally used a molding sand that had been burned and mixed different amounts of sea coal with it.

(Mr. Grennan then discussed the tables of the paper.)

* University of Michigan, Ann Arbor, Mich.

CHAIRMAN R. F. HARRINGTON: We appreciate Mr. Grennan's effort in abstracting this paper. Mr. Hird is one of our ablest investigators on this subject of sea coal, and I am sure we all agree that it is a vital subject in every foundry and one concerning which we can probably look forward to a good deal of conflict in ideas.

MEMBER: Was the 15 per cent of the sea coal mentioned, the percentage by volume or by weight?

JOHN GRENNAN: There is nothing in this paper that would indicate whether it is by volume or by weight. I do not think there is much difference between the volume and the weight, if one measures them. Carbon has a specific gravity of about 2, and the sand is higher than that; thus, there would be some difference.

PAUL BECHTNER: I would say that a scoopful of sea coal weighs much less than a half scoopful of sand, so there would be a much greater difference if it was volume rather than weight.

Did Mr. Hird mention the effect of a reducing atmosphere created by sea coal?

JOHN GRENNAN: The author does not go into that question, but making a guess on my own initiative, I believe that the atmosphere in a mold is a reducing atmosphere. If it was an oxidizing atmosphere, the soots and gases would be burned up.

PAUL BECHTNER: That is so, but a few years ago Mr. Goodale published a paper² in which he showed that an oxidizing atmosphere on the surface led to the formation of silicates that were highly fusible, whereas the reducing atmosphere did not do that. Drawing an analogy from ceramics, we have seen sewer-pipe plants where they maintain in a kiln what they call a water-smoking period during which the coal is kept damp and a heavy smoke is carried through the kiln. The reason for that is that there are particles of iron in the clay which, unless they are reduced and made to flow, will make black spots on the surface of the sewer pipe, resulting in a defect, and the reducing atmosphere overcomes that.

It has been a theory of mine, without any real scientific basis, that a large part of the advantage of the sea coal was due to the reducing atmosphere that it created, or the additional reducing atmosphere—if there is some originally, and there probably is—making the iron more fluid.

H. W. DIETERT: The table shown begins, I believe, with 15 per cent sea coal, and the sea coal was reduced or burned out. The casting as shown had "grapevines," indicating that this 15 per cent of sea coal was entirely too high. I am inclined to believe that is the reason why the strength was low at the beginning. I have never seen any American sea coal that increases the green-sand strength.

JOHN GRENNAN: The author makes the statement quite definitely in his paper that hand mixing does not bring out the influence of the sea coal, whereas mixing with a muller, or a pan mix, brings out the bonding property of the sea coal.

* Hunt-Spiller Mfg. Corp., Boston.

† American Colloid Co., Chicago.

² Goodale, P. L., *Notes on the Behavior of Sand Molds in Steel Foundries*. TRANSACTIONS A.F.A. (1930), v. 38, pp. 471-478.

** U. S. Radiator Corp., Detroit.

Indirectly, I checked up the volatile in the sea coal such as we use in the United States. Wisconsin County Coke will run about 35 or 30 per cent volatile, and that kind of a coal in a coke oven at 500 degrees distills off something like 9½ per cent tar. At 1000 degrees, the tar drops to about 5 per cent but the density of the tar increases, so that it is a more viscous material. It is common practice in the United States to use the tar products in the form of pitch binders in cores. Consequently, I believe the author has considerable backing for his assertion that there is bonding property in the volatile of the sea coal.

Martensitic Quenching of Cast Iron*

With Special Reference to Low-Carbon Irons With High Manganese Contents

By GEORGES R. DELBART,† D.Sc., DENAIN, FRANCE

Abstract

Investigations have been made, following the work of Guillet, Galibour and Ballay, on the martensitic quenching of alloy cast irons. A number of experiments were made particularly with a view to studying the suitability for quenching of irons having a low content of total carbon and a high manganese content. It was found that manganese, as soon as it exceeds 2 per cent, definitely favors the martensitic quenching to such an extent that medium-size pieces with a thickness attaining 1.06 in. (27 mm.) could take a martensitic structure simply by cooling in still air. On the other hand, the low carbon content has improved the mechanical properties and has considerably reduced the danger of microscopic cracks or fissures due to rapid cooling, and this notwithstanding the presence of a high manganese content. These proportions of low total carbon and high manganese have contributed to the increased stability of irons subjected to heat, a property to be looked for in irons to be subjected to heat treating. Very low contents in total carbon should be avoided in current practice, owing to the decrease of the castability, to the rapidity with which the metal passes from the liquid to the pasty state, and to its tendency to piping. A content of about 2.50 per cent can be adopted when melting iron in the cupola; in the electric furnace, which enables the iron to be superheated, interesting mechanical properties should be obtained with higher carbon, which can be around 2.80 per cent. Finally, if the investigations of Guillet, Galibour and Ballay are considered, interesting combinations are possible that can be used in practice in preparation of nickel-manganese irons with low total carbon and high manganese content. Similarly, one could expect to see the nickel neutralize the action of the manganese, which in any case is slight, on the primary quenching (hardening), while their effect on the reaction to martensitic quenching would be

*Exchange paper of the French Foundry Technical Association, 1933.

†Ateliers des Anciens Etablissements Cail.

NOTE: This paper was presented before one of the gray iron sessions at the 1933 Convention of American Foundrymen's Association.

added to each other, resulting in irons having a particular tendency for quenching. Furthermore, the low carbon contents, or else superheating, would tend to decrease the danger of microscopic cracks or fissures, a danger which the author has not encountered during investigations on pieces having a simple shape, the results of the mechanical tests not having indicated any depreciation in the qualities of the iron, but rather the reverse.

1. Modern requirements often call for the use of hard gray irons with considerable resistance to friction and abrasion. On the other hand, manufacturers require that such irons should be machinable, a qualification which often eliminates the use of irons containing a large percentage of cementite, such irons being hard and at the same time brittle and very difficult to machine.

2. Primary hardening, which consists of casting in chilled molds in order to accelerate the cooling rate at the time of solidification, and which permits of obtaining castings which present on the outside surface a more or less thick layer of white iron, can be applied only in very restricted fields, as in the manufacture of rolls for plate and sheet mills. It therefore appears practical to apply to machinable gray irons the secondary or martensitic quenching, such as is currently practiced in order to improve the mechanical properties of steel. However, in this case the problem is complicated by the presence of graphite and by the phenomenon of graphitization, which is liable to appear during the heating period necessary to bring the iron to its quenching temperature.

3. Portevin, in his study on the heat treating of cast-iron shells, showed, in 1922, that the rapid cooling of so-called semi-steel often caused porosities to appear, due to microscopic cracks. It therefore is dangerous to quench, by rapid cooling, ordinary irons which have high critical rates of quenching. Consequently, it is of interest to add certain special elements which tend to reduce the critical rate of quenching to that which corresponds to the cooling in still air of the castings to be treated.

4. It is known, on the other hand, that most ordinary cast irons give rise during heating to graphitization, which is accompanied by a permanent expansion and by the formation of internal stresses.

5. This graphitization occasionally begins at low temperatures—about 932 degs. Fahr. (500 degs. Cent.)—and in certain conditions it is sufficiently rapid for the graphitization of the pearlite to

be completed in a few hours. This results in a decrease in the reaction to quenching, a lowering of the mechanical properties, and the appearance of internal stresses that are unfavorable to heat treatment. These internal stresses are greater as graphitization takes place at lower temperatures.

6. Briefly, an ideal cast iron for martensitic quenching should answer to the following conditions: (1) The graphite content should be as low as possible; (2) the graphite should be fine in texture and rounded in shape; (3) the chemical composition of the iron should be such that the iron is stable, and the temperature of graphitization should be higher than the temperature of transformation; (4) the critical rate of quenching should be as low as possible.

7. The first attempts to produce irons answering to the fourth condition (which is the most important) were made, to the author's knowledge, by Guillet, Galibour and Ballay, and were given in the *Comptes-rendus de l'Academie des Sciences* in 1928, and in the *Revue de Metallurgie*, November, 1931. Subsequently, in collaboration with Edgar Lecoeuvre, we have produced in the cupola a certain number of irons with low carbon content which have shown themselves to be particularly suited to martensitic quenching.

8. Before reviewing the results obtained, we wish to review certain characteristics of the principal elements which are liable to enter into the chemical composition of irons suitable for martensitic quenching.

Silicon

9. Silicon favors graphitization, increases the critical transformation range, increases the critical rate of quenching, and decreases the capacity for martensitic quenching.

Manganese

Influence on Graphitization.

10. Manganese decreases the tendency to graphitization and raises the critical temperature under which graphitization takes place. Table 1 gives examples from the author's personal investigations on irons having a low carbon content.

11. In Table 1 the values given for expansion correspond to the expansion obtained after two cycles of heating at 1652 and 1832 degs. Fahr. (900 and 1000 degs. Cent.) The heating rate was 32.4 degs. Fahr. (18 degs. Cent.) per minute, and the rate of cool-

Table 1

TC*	Si	Mn	Critical Temperatures, degs. Fahr.	Expansion	Remarks
2.1	3.3	2.14	1382	2.2×10^{-3}	High Si content.
2.08	2.95	0.45	1247	5.2×10^{-3}	High Si content.
2.0	2.26	1.54	1382	1.1×10^{-3}	Medium Si content.
2.14	2.16	0.48	1310	3.7×10^{-3}	Medium Si content.
2.5	3.2	2.12	1400	2.2×10^{-3}	High Si content.
2.42	3.02	0.49	1238	8.5×10^{-3}	High Si content.
2.4	2.5	1.43	1409	1.1×10^{-3}	Medium Si content.
2.5	2.4	0.55	1328	4.5×10^{-3}	Medium Si content.

*The first 4 irons were made to contain 2 per cent Total Carbon; the second 4 irons, 2.5 per cent Total Carbon.

Table 2

TC	Si	Mn	Ac		Ar		Remarks
			Beginning, °F.	End, °F.	Beginning, °F.	End, °F.	
2.00	2.95	0.47	1490	1562	1310	1130	
1.97	2.83	2.04	1427	1472	266	...	
2.21	2.86	4.0	1382	1454	248	...	
2.24	2.89	4.7	1382	1454	248	...	
2.2	2.28	0.70	1472	1526	1274	1094	
2.0	2.26	1.54	1463	1499	1184	1004	All specimens were cooled in still air. The first 4 specimens were considered high Si; the last 2, medium Si.

ing was 47 degs. Fahr. (26 degs. Cent.) per minute. All the irons on which investigations were made were prepared under identical conditions, and the test pieces used for dilatometric investigations were taken from blocks of the same thickness. Therefore, the thermal history of the test pieces should not have to be taken into consideration to explain the differences found in the results.

Influence on Capacity for Quenching.

12. Table 2 concerns cast irons with 2 per cent total carbon and varying contents of manganese, and the examples given are suggestive.

13. From these results it can be concluded that manganese lowers the transformation points during the heating cycle, particularly when the manganese contained reaches 2 per cent; but the lowering of the transformation point during the cooling cycle is already well marked for 1.5 per cent manganese. However, the lowering of the points during the cooling cycle is still more marked when the content of manganese reaches 2 per cent, at which point the test pieces are susceptible to martensitic quenching by simply cooling in still air.

14. Guillet, Galibour and Ballay have shown that, with a content of 1 per cent manganese and 1 to 2 per cent nickel, a

marked action on the effect of quenching is noticeable, because the influence of manganese and nickel is in the same direction and their respective influences are added to each other. On the other hand, manganese has only a slight effect on primary hardening, and its action under 2 per cent is negligible, at least when the silicon content is normal. Beyond 2 per cent, the action of manganese is slightly more marked and should be neutralized by a slight increase in the amount of silicon. Manganese increases hardness and reduces the deflection in the transverse test as soon as the percentage reaches and exceeds 2 per cent. Manganese tends to decrease the heterogeneity of the 2nd degree.*

Table 3

Iron No.	TC	GC	CC	Si	Mn	P	S	Ni
36	2.45	1.58	0.87	1.39	0.54	0.340	0.007	2.05
5	2.47	1.40	1.07	1.60	0.44	0.356	0.090
19	1.90	1.10	0.80	2.70	0.60	0.110	0.070
29	1.72	0.93	0.79	2.90	2.50	0.072	0.051

Nickel

15. Nickel lowers the critical temperature range and the critical rate of quenching, and therefore acts very favorably on the secondary quenching. It has the advantage over manganese of reducing the tendency to primary hardening. It favors graphitization, but this action can be neutralized by lowering the percentage of silicon and by slight additions of chromium. Nickel markedly decreases the heterogeneity of the 2nd degree.

Chromium

16. Chromium raises the critical temperature range but lowers the critical rate of quenching. It favors an increase in the depth of hardening. It raises the limit of the temperature of reversibility and tends to decrease graphitization.

Carbon

17. The higher the content of carbon, the greater the tendency toward graphitization and the greater the heterogeneity of the 1st and 2nd degrees. The graphite is all the finer as the content of total carbon is lower. In cast irons with 2.5 to 2.8 per

*For the definition of heterogeneity of the 1st and 2nd degrees, see "Les Fontes à Haute Resistance," Lemoine, *Liege Congress*, 1928.

cent total carbon, made in the cupola, eutectic is found in thin test pieces of 2.95 in. (75 mm.) thickness and which, therefore, have been cooled rapidly; whereas, in pieces 5.90 in. (150 mm.) thick, the graphite is more readily found to be in the form of lamellae.

18. However, in gray irons having relatively little tendency to graphitization, in irons having a low total carbon, and in irons with a low percentage of silicon situated in the pearlitic region of the Maurer diagram but at the limit of mottled iron, the graphite tends to precipitate also in the 5.90-in. (150 mm.) pieces in the form of fine eutectic. We have had some very marked examples of irons having the analysis shown in Table 3, in which the structural heterogeneity was, therefore, very small (see Figs. 1-5, inclusive).

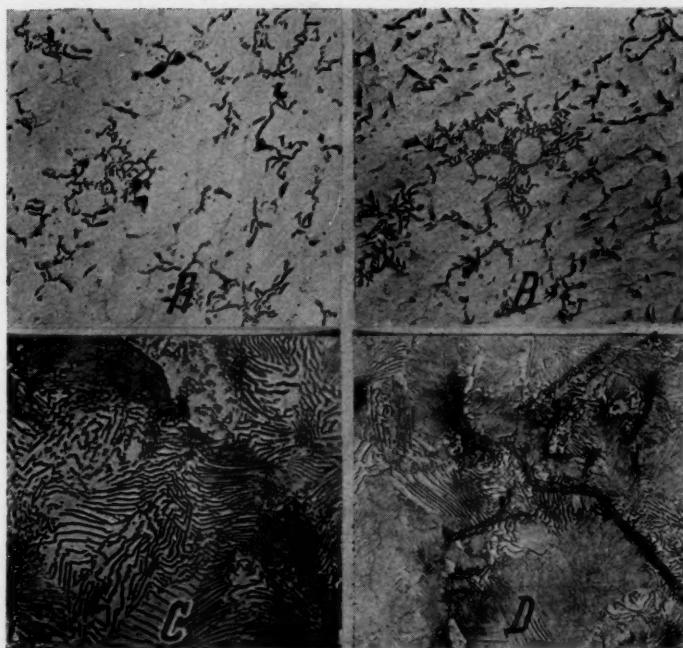


FIG. 1—PHOTOMICROGRAPHS OF IRON NO. 36 (TC 2.45 PER CENT, GC 0.34, SI 1.39, Mn 0.54, Ni 2.05). A: 150 MM. TEST PIECE, UNETCHED, X50. B: 15 MM. TEST PIECE, UNETCHED, X50. C: 150 MM. TEST PIECE, ETCHED WITH PICRIC ACID, X400. D: 15 MM. TEST PIECE, ETCHED WITH PICRIC ACID, X400.

STUDY OF NICKEL CAST IRONS

19. Guillet, Galibour and Ballay have studied a comparatively large number of various irons made in the crucible with compositions within these limits: Total carbon 3 to 3.4 per cent, manganese 0.24 to 1.32 per cent, silicon 0.84 to 2.11 per cent, nickel 0 to 5.4 per cent, chromium 0 to 0.7 per cent, sulphur under 0.06 per cent, phosphorus 0.37 to 0.96 per cent. From the practical point of view they considered two groups: (1) Irons easily machined and susceptible to martensitic hardening; (2) self-hardening irons which are hard when cast but are susceptible to machining after having been reheated at 1202 degs. Fahr. (650 degs. Cent.)

20. Further investigations were directed principally toward the following points: (a) The determination of the critical points

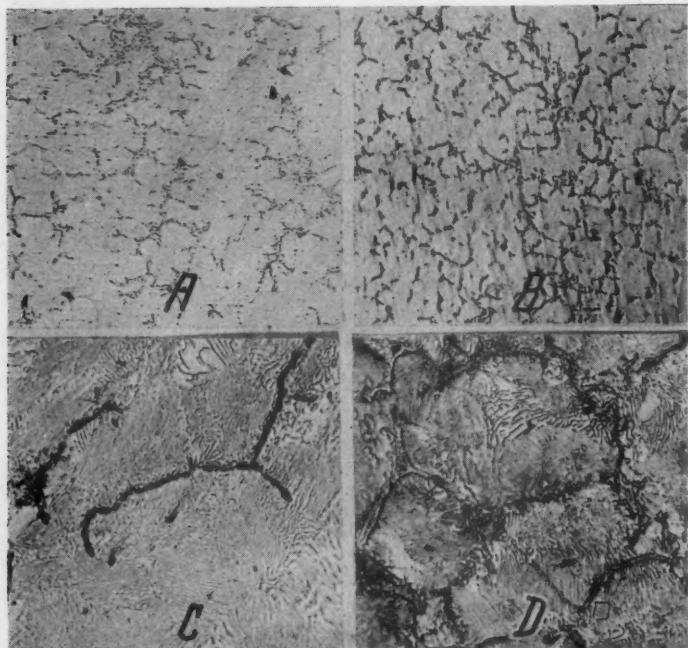


FIG. 2—PHOTOMICROGRAPHS OF IRON NO. 5 (TC 2.47 PER CENT, GC 0.356, Si 1.60, Mn 0.44). A: 150 MM. TEST PIECE, UNETCHED, X50. B: 15 MM. TEST PIECE, UNETCHED, X50. C: 150 MM. TEST PIECE, ETCHED WITH PICRIC ACID, X400. D: 15 MM. TEST PIECE, ETCHED WITH PICRIC ACID, X400.

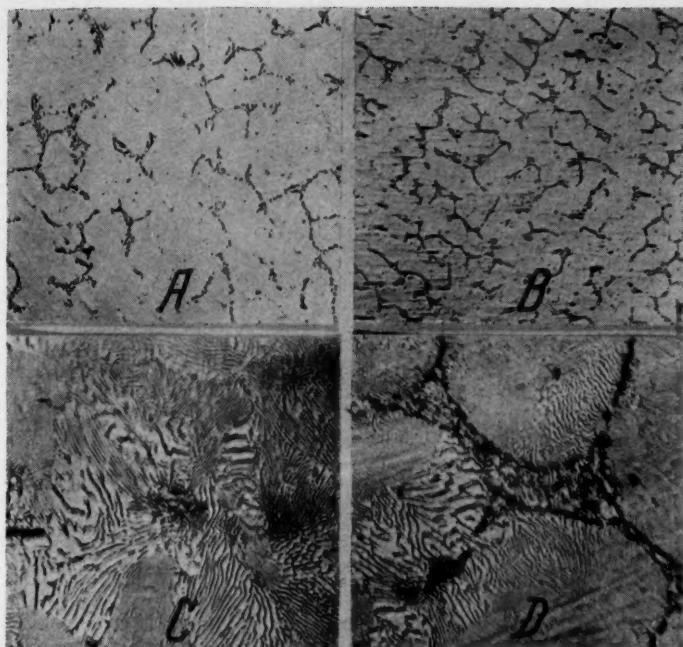


FIG. 3—PHOTOMICROGRAPHS OF IRON NO. 19 (TC 1.90 PER CENT, SI 2.7, MN 0.60). A: 150 MM. TEST PIECE, UNETCHED, X50. B: 15 MM. TEST PIECE, UNETCHED, X50. C: 150 MM. TEST PIECE, ETCHED WITH PICRIC ACID, X400. D: 15 MM. TEST PIECE, ETCHED WITH PICRIC ACID, X400.

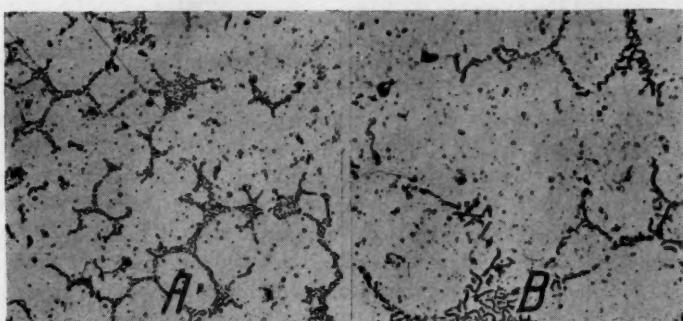


FIG. 4—PHOTOMICROGRAPHS OF IRON NO. 29, TEST PIECES 15 MM. THICK. UNETCHED, X80. A: IN THE GRAY PORTION, THE GRAPHITIC EUTECTIC HAS TAKEN THE PLACE OF CEMENTITE. B: IN THE CENTER, THE GRAPHITIC EUTECTIC IS STILL FINE.

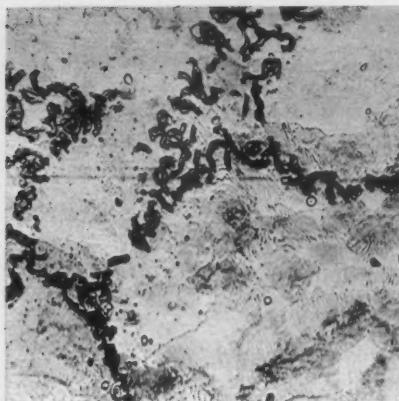


FIG. 5—PHOTOMICROGRAPH OF IRON NO. 29, TEST PIECE 15 MM. THICK, STRUCTURE IN THE GRAY PORTION. MATRIX IS ENTIRELY PEARLITIC. TOTAL ABSENCE OF FERRITE, EVEN IN PROXIMITY OF GRAPHITE. PEARLITE IS IN FINE GRAINS. SOME PARTICLES OF CEMENTITE. ETCHED WITH PICRIC ACID, X1000 (REDUCED APPROXIMATELY ONE-FOURTH).

for various rates of cooling; (b) determination of hardness curves as functions of quenching temperatures; (c) the influence of drawing; (d) mechanical properties after quenching and drawing.

21. The principal conclusions drawn were that it is possible to obtain irons which harden by martensitic quenching with a relatively small content of nickel, from 1 to 2 per cent, provided the content of silicon is rather low and the manganese content sufficient; that the manganese should be about 1 per cent in order to facilitate the influence of nickel. Generally, silicon can be maintained at about 1 per cent.

22. For quenching in oil, a nickel content of 1.5 to 2 per cent is sufficient. The analysis adopted for test pieces 0.787 in. (20 mm.) thickness was: Total carbon 3.2 per cent, manganese 1 per cent, silicon 1.1 to 1.15 per cent, and nickel 1.5 to 2 per cent.

23. For quenching in still air, the nickel content should be raised to 4 or 5 per cent, according to the thickness. As the pieces must be hard when cast, they must have a relatively high content of nickel because the cooling rate in sand is low. Tables 4 to 10, inclusive, give a few results as indicated by the investigators.

24. Finally, the investigators have certain data on machinability in a lathe which showed that the irons under consideration can generally be easily machined.

Table 4

Iron No.	Cooling Rate.	TC	Mn	Si	Ni	P	S	Ac		Ar ₁	
								Begin- ning °F.	End, °F.	Begin- ning °F.	End, °F.
175	About 11 min.	3.38	0.57	2.00	0.00	0.37	0.062	1490	1571	1469	1310
176	About 11 min.	3.12	0.60	1.46	1.15	0.39	0.065	1427	1517	1358	1265
252	About 11 min.	3.35	1.13	1.73	1.82	0.73	0.060	1427	1517	1274	1184
205	About 11 min.	3.42	1.10	1.80	3.32	0.19	0.141	1355	1454	1103	1040
205	About 30 ² ° F. per min.	3.42	1.10	1.80	3.32	0.19	0.141	1355	1454	1184*	437*
215	About 20° F. per min.	3.08	0.92	1.57	4.02	0.08	0.079	1292	1418	392
216	About 20° F. per min.	3.07	0.91	1.70	5.07	0.08	0.077	1247	1427	347

*Only slightly indicated.

Table 5

PHYSICAL PROPERTIES—IRON NO. 176.

Brinell Hardness as cast = 188
(Specimen 1.77 in. (45 mm.) in diameter)

Quenching Temperature, degs. Fahr.	Brinell Hardness	
	Oil Quenched	Air Quenched
1382	176	172
1472	265	186
1562	325	201
1652	389	223

BRINELL HARDNESS AFTER QUENCHING IN OIL
AT 1652 DEGS. FAHR. AND DRAWN

Temperature of Draw, degs. Fahr.....	68	482	572	797	932	1202	1337
Brinell Hardness number.....	389	339	307	315	271	178	154

Table 6

IRON NO. 252—PROPERTIES AS CAST AND AFTER OIL QUENCHING
AT 1490 DEGS. FAHR. AND DRAWN.

As Cast.	Quenched and Drawn at—				
	68° F.	572° F.	797° F.	842° F.	1112° F.
Brinell Hardness number...	170	353	447	290	309
Compression, lbs. per sq. in.	113,760	107,076.6	163,530	183,576
Shear,* lbs. per sq. in.	36,261	28,724	42,944.4	46,926	44,793
Transverse,* lbs.	1135	540	1124	1212
Deflection, mm.	0.17	0.14	0.14	0.15
					0.18

*Shearing tests were made on cylindrical pieces of 0.22 in. (5.64 mm.) diameter, and the transverse tests with pieces 0.315 x 0.394 in. (8 x 10 mm.) on supports 1.18-in. (30 mm.) apart as per Specifications of French Navy.

Table 7

IRON NO. 205—BRINELL HARDNESS OF TEST PIECES AS CAST.

Specimen 0.98 in. (25 mm.) in Diameter	Specimen Diam. 1.77 in. (45 mm.) At the Surface	At the Center
254	269	225

Table 8

BRINELL HARDNESS AFTER QUENCHING IN OIL AND AIR BLAST AND DRAW

Temperature of Draw, degs. Fahr.....	68	482	572	662	797	932	1067	1202	1337	1517
Brinell Hardness No. when Quenched in Oil.....	434	398	374	365	371	313	299	263	242	337
Brinell Hardness No. when Quenched in Air Blast.....	389	303	322	317	325	262	250	198	196	297

Table 9
IRON No. 216

Temperature of Quench, degs. Fahr.	Hardness After Quenching		
	In Oil	In Air Blast	In Still Air
1292	386	305	...
1382	371	401	...
1472	347	347	...
1562	322	322	...
1652	301	301	332

Table 10
IRON No. 216

Quenched in Oil at 1292 degs. Fahr. and Drawn at:

Degs. Fahr.....	68	662	797	1067	1202	1337	1517
Brinell Hardness No.....	386	317	285	258	218	214	315

25. In 1931, Raymond Chavy of the Centre of Information on Nickel presented a paper to the International Foundry Congress at Milan entitled "Quenched and Hyper-Quenched Cast Irons—Investigations and Recent Applications," in which he examined the following from the point of view of practical applications and results: (a) White irons with cementite-martensite; (b) pearlitic gray irons or sorbitic irons; (c) gray martensitic irons, after cooling in the mold, and after cooling in still air following immediate stripping from the mold; (d) gray irons with martensite-austenite; (e) austenitic gray irons. The study of this last class of irons, which had been started in the United States, was taken up and developed in France by Marcel Ballay. These austenitic irons present considerable interest from the point of view of their resistance to corrosion, and their electro magnetism.

26. In the studies summarized above, it is more particularly the influence of nickel that was investigated, bearing in mind that the favorable influence of a manganese content reaching 1 to 1.2 per cent had already been found. The action of manganese on the quenching properties of steel has been known for a long time,

Table 11

Carbon, Per Cent	Manganese, Per Cent	Structure
0.80	3.0	Pearlitic.
	3 to 7	Martensite-Troostite.
1.00	1.9 to 2.0	Martensite-Troostite.
	2.5 to 4.8	Martensite.

and from the structural diagram of manganese steels by Guillet it is apparent that for contents ranging from 0.8 to 1 per cent of combined carbon, which correspond to the contents of pearlitic irons, the structures of Table 11 are obtained.

27. On the other hand, Portevin and Chevenard have touched upon the study of the influence of manganese in cast iron, and they gave in the bulletin of the Association Technique de Fonderie some dilatometric curves relative to an iron having the composition of carbon 4.2 per cent, silicon 1.34 per cent, and manganese 4.06 per cent. They summarized the action of the manganese as follows: (1) Manganese tends to neutralize graphitization; (2) it has an action on the constitution of complex cementites; (3) it renders iron-carbon alloys self-hardening (air quenching).

28. We believe that these influences of manganese have not been commercially applied up to the present, for the reason that it was feared the manganese had a strong influence on the primary hardening, and because manganese was, and still is, believed to facilitate the formation of porosities and microscopic cracks.

29. We ourselves have had experience with manganese irons having a coarse grain structure, which are occasionally used and made in America as well as in Europe. The most current practical formula for these irons with coarse grain structures is as follows: Total carbon 3.4 per cent, silicon 3 per cent, manganese 3 per cent, and phosphorus 0.3 to 0.6 per cent. However, we also have found similar crystalline fractures in irons having the same chemical composition, except that the manganese was under 2 per cent, and it is quite certain that if the manganese has an influence on the constitution of these irons, the total carbon, silicon and phosphorus are no less important factors of heterogeneity. We show in Fig. 6 an example of the fracture of such an iron, and in Fig. 7 a microstructure of such an iron. Such irons have mediocre mechanical properties and are utilized only on account of their coarse grain structures.

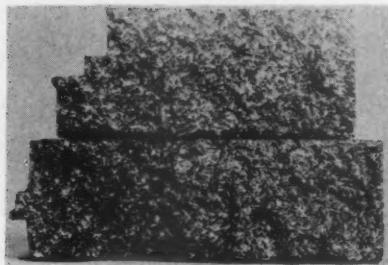


FIG. 6—APPEARANCE OF FRACTURE OF IRONS NOS. 40 AND 43. REDUCED APPROXIMATELY ONE-FOURTH.

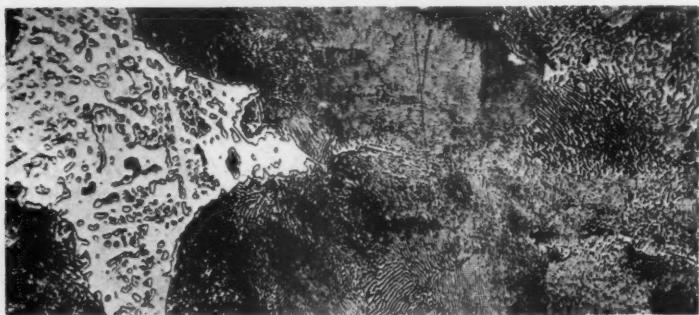


FIG. 7—PHOTOMICROGRAPH OF IRON NO. 37, STRUCTURE SHOWING GRAPHITE, PEARLITE AND PHOSPHORUS EUTECTIC. ETCHED WITH NITRIC ACID, X400 (REDUCED APPROXIMATELY ONE-FOURTH).

30. From these results we have concluded that it is, no doubt, possible to obtain manganese irons with a fine structure but possessing high mechanical properties, by varying in a favorable direction the elements other than manganese. The practical results have come up to our expectations, and after lowering the contents of carbon and phosphorus we have obtained homogeneous irons with very fine structure and susceptible to martensitic quenching.

31. We are well aware that the disadvantage of low-carbon cast irons is that they are hard to cast and that they are more susceptible to piping; but it is possible in common practice, even with an ordinary cupola, to make cast irons of a suitable chemical composition in which the carbon is fixed, for example, at a minimum of 2.3 per cent. What is possible in the cupola is still more possible in the reverberatory furnace and in the electric furnace.

Table 12

Iron No.	TC	GC	CC	Si	Mn	P	S	Cr	Tensile Strength, ¹ lbs. per sq. in.	Brinell Hardness No.	Heterogeneity of the 2nd degree, in %.
IRONS WITH LOW CARBON AND MANGANESE CONTENT											
14	2.08	1.52	0.74	2.95	0.47	0.084	0.079	47,780
15	2.20	1.44	0.78	2.90	0.47	0.080	0.067	56,880
21	2.50	1.71	0.89	2.40	0.55	0.210	0.082	45,500	269	10.4
24	2.24	1.50	0.74	2.28	0.70	0.074	0.075	44,080	269	15.5
IRONS WITH LOW CARBON AND HIGH MANGANESE CONTENT											
25	2.40	1.50	0.90	2.50	1.43	0.156	0.075	46,357	275	7.95
26	2.00	1.46	0.54	2.26	1.54	0.088	0.072	47,495	277	21
27	2.11	1.47	0.64	3.35	2.14	0.148	0.033	43,371	340	13.7
28	2.50	1.62	0.83	3.22	2.12	0.280	0.059	46,215	293	8
29	1.72	0.93	0.79	2.90	2.50	0.072	0.051	44,508	335	9.3
30	1.97	1.09	0.88	2.85	2.06	0.102	0.048	48,206	321	11.1
31	1.69	1.10	0.59	2.89	3.75	0.118	0.026	286 ²	nil
31-B	2.21	1.59	0.62	2.86	4.0	0.120	0.038	311
31-C	2.24	1.49	0.75	2.89	4.7	0.100	0.034	321
IRONS WITH LOW CARBON CONTENT CONTAINING MANGANESE AND CHROMIUM											
44	2.00	1.10	0.90	2.28	3.27	0.172	0.033	0.32	41,950	340	17
45	2.27	1.61	0.66	4.52	2.35	0.108	0.056	0.47	43,371	302	13.9
46	2.23	1.47	0.76	3.78	2.60	0.154	0.057	0.48	322	14.1
IRONS WITH HIGH CARBON, SILICON AND MANGANESE CONTENT											
37	3.42	2.74	0.68	2.81	2.39	0.316	0.020	22,468	207
38	3.32	2.69	0.63	2.00	2.24	0.268	0.013	20,761	207
39	3.22	2.71	0.51	2.77	2.01	0.478	0.017	24,032	207
40	3.15	2.46	0.69	3.07	1.77	0.420	0.017	20,477
41	3.21	2.57	0.64	3.05	1.67	0.442	0.025	24,600
42	2.96	2.24	0.72	2.88	1.43	0.454	0.026	21,900	196
43	2.90	2.25	0.65	2.73	1.29	0.422	0.020	25,170

¹Tensile tests made on specimens with an "as cast" diameter of 1.18 in. (30 mm.) and machined to a diameter of 0.98 in. (25 mm.)

²Hardness taken in a block of 9.84 in. (250 mm.) diameter and 12.78 in. (325 mm.) high; hardness in the center and at the surface is the same; the heterogeneity of the first degree is, therefore, nil.

Table 13

Iron No.	Brinell Hardness No.	Shear Strength, lbs. per sq. in.	Transverse Strength, lbs. per sq. in.	Deflection, mm.	Compression, lbs. per sq. in.	Remarks
14	340	56,880	2205	0.20	184,860	{ The test pieces were taken in a square bar 1.58 x 1.58 in. (40 x 40 mm.)
15	340	53,325	1996	0.18	209,034	

32. In Table 12 are given the characteristics of a number of irons with low carbon content which we have studied, and more particularly those samples having a high content of manganese.

33. The tensile strength was determined on test pieces having a diameter of 1.18 in. (30 mm.) when cast, and machined down to 0.98 in. (25 mm.) For the irons of Table 13 the Brinell hardness was determined on pieces 1.57 x 1.57 in. (40 x 40 mm.), and on the irons of Table 12 the heterogeneity of the 2nd degree was

Table 14
TESTS ON STEP BAR SPECIMENS, SHOWING RELATION OF HARDNESS, SHEAR STRENGTH, TRANSVERSE STRENGTH AND DEFLECTION OF IRONS OF FOUR DIFFERENT SECTION THICKNESSES.

Iron No.	-0.39 in. (15 mm.) Section			-1.08 in. (30 mm.) Section			-2.16 in. (60 mm.) Section			-3.54 in. (150 mm.) Section		
	Brinell Strength, lbs. per sq. in.	Shear Strength, transverse, lbs. per sq. in.	Deflection, mm.	Brinell Strength, vers., lbs. per sq. in.	Shear Strength, vers., lbs. per sq. in.	Deflection, mm.	Brinell Strength, trans., lbs. per sq. in.	Shear Strength, trans., lbs. per sq. in.	Deflection, mm.	Brinell Strength, trans., lbs. per sq. in.	Shear Strength, trans., lbs. per sq. in.	Deflection, mm.
21	266	55,458	1797	0.15	255	54,036	2017	0.16	248	51,192	1982	0.19
24	286	45,788	1830	0.12	269	55,458	1929	0.16	262	52,330	1764	0.14
25	277	60,493	1991	0.14	269	55,458	1929	0.16	262	52,330	1764	0.14
26	302	52,898	2028	0.11	293	55,458	2017	0.16	260	52,330	1764	0.14
27	321	52,898	1411	0.10	311	55,458	2017	0.16	286	52,330	1764	0.14
28	302	48,916	1708	0.12	286	48,348	1632	0.12	286	48,348	1543	0.11
29	340	61,000	1841	0.12	335	54,036	1676	0.11	332	56,018	1764	0.12
30	332	52,898	1797	0.11	321	53,467	1962	0.11	302	52,330	1764	0.12
44	364	49,585	1279	0.06	340	55,458	1929	0.06	340	52,330	1764	0.06
45	321	38,394	882	0.10	302	55,458	1929	0.06	302	52,330	1764	0.10
46	340	47,779	1279	0.08	322	55,458	1929	0.08	322	52,330	1764	0.08

determined from hardness tests made on step-bar test pieces in which the thicknesses ranged from 0.59 to 5.9 in. (15 to 150 mm.)

34. The mechanical properties of some of the above-mentioned irons, tested according to the specifications of the French navy, are given in Table 14. The tests were made on test pieces of which the dimensions are indicated above, with a Malicet and Blin testing machine. As a basis of comparison, we might say that the provisional specifications of the French navy for high-strength cast iron are: Brinell hardness 170 to 240, shear strength 35,550 lbs. per sq. in., transverse strength 1213 lbs., deflection 0.2 mm., and compression 110,916 lbs. per sq. in. Table 14 gives the results of tests on step-bar test pieces.

35. Irons Nos. 31-B and 31-C were tested on pieces taken

Table 15

Iron No.	Brinell Hardness No.	Shear Strength, lbs. per sq. in.	Transverse Strength, lbs.	Deflection, mm.
31-B	306	54,885	2205	0.12
31-C	340	56,880	1984	0.11

from a cylinder of 9.8 in. (250 mm.) diameter, with a thickness of 0.79 in. (20 mm.) The results were as given in Table 15.

36. The conclusions to be gained from these tests have already been largely given in the paragraph in which we discussed the influence of manganese. One can note particularly the reduction of the deflection as the manganese content and the hardness increase. The transverse strength is considerably reduced in the manganese-chromium irons Nos. 44 to 46, and it is particularly low, as is also the shearing strength, in the iron No. 45 with 4.52 per cent silicon. Generally speaking, the mechanical characteristics are much higher than the specifications of the French navy.

MICROSTRUCTURE

37. The graphite in the irons investigated is never surrounded by ferrite, and even the graphitic eutectic contains pearlite in the space between the lamellae. We have seen in the paragraph relating to carbon that the structural heterogeneity is very small in irons which have little tendency to graphitization, and that the graphite was found in the eutectic state in the thick test pieces as well as in the thin ones. In the manganese irons, an example

of which is given by iron No. 29 (Figs. 4 and 5), the irons containing 2.2 to 5 per cent manganese were distinctly pearlitic, and in the iron containing 4 per cent manganese there is, in addition, a little martensite in the pieces having a thickness of 0.787 in. (20 mm.)—see Figs. 8 and 9. A simple reheating under the point of transformation (Ac_1) is sufficient to destroy this martensite and to soften the iron.

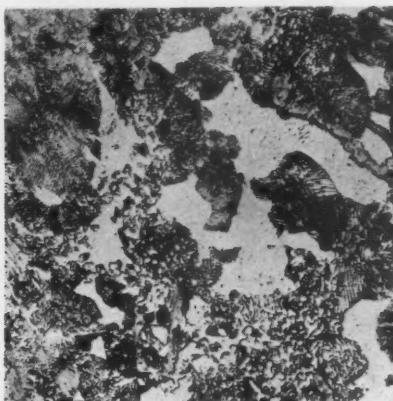


FIG. 8—PHOTOMICROGRAPH OF IRON WITH 4.73 PER CENT MANGANESE (TC 2.25 PER CENT, SI 2.89), STRUCTURE SHOWING PEARLITE, GRAPHITE AND MARTENSITE. ETCHED WITH PICRIC ACID, X1000 (REDUCED APPROXIMATELY ONE-FOURTH).

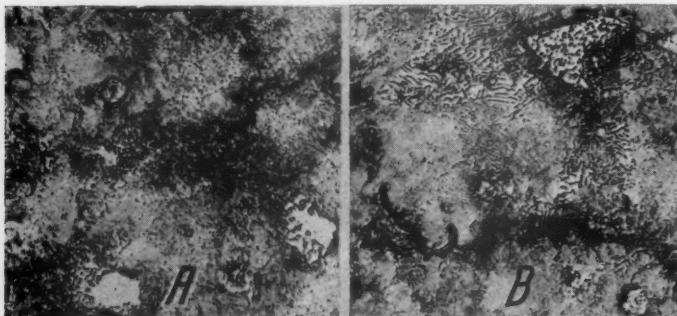


FIG. 9—PHOTOMICROGRAPHS OF HIGH-MANGANESE HEAT-TREATED IRONS, DRAWN AT 700 DEGS. CENT. (1292 DEGS. FAHR.) FOR 15 MINUTES, AND COOLED IN FURNACE. ETCHED WITH PICRIC ACID, X300. A: IRON WITH MANGANESE 4.75 PER CENT, TC 2.25, SI 2.89, SHOWING GRAPHITE, PEARLITE AND SPOTS OF CEMENTITE OF FERROALLOY NOT COMPLETELY DISSOLVED. B: IRON WITH MANGANESE 3.99 PER CENT, TC 2.21, SI 2.85, STRUCTURE SHOWING GRAPHITE AND PEARLITE.

Table 16

Iron No.	TC	Si	Mn	Ac		Ar		Critical Temp., °F.	Expansion.	Ac-Ar °F.
				Beginning, °F.	End, °F.	Beginning, °F.	End, °F.			
14	2.06	2.95	0.47	1490	1562	1310	1130	1247	5.2 x 10^-3	284
15	2.20	2.90	0.47	1490	1562	1310	1130	1256	5.3 x 10^-3	284
21	2.50	2.40	0.55	1472	1544	1310	1130	1328	4.5 x 10^-3	266
24	2.24	2.28	0.70	1472	1526	1274	1094	1337	3.5 x 10^-3	284
25	2.40	2.50	1.43	1436	1544	1238	1022	1409	2.45 x 10^-3	338
26	2.00	2.26	1.54	1463	1499	1184	1004	1382	1.1 x 10^-3	347
27	2.11	3.35	2.14	1490	1526	320	...	1382	2.2 x 10^-3	1238
28	2.50	3.22	2.12	1490	1526	302	...	1400	2.2 x 10^-3	1256
29	1.72	2.90	2.50	1436	1472	257	...	1682	0.7 x 10^-3	1247
30	1.97	2.85	2.06	1427	1472	266	...	1472	1.35 x 10^-3	1238
31	1.60	2.89	3.75	1382	1472	167	...	1697	1.25 x 10^-3	1337
31-B	2.21	2.86	4.00	1382	1454	248	1238
31-C	2.24	2.89	4.70	1382	1464	248	1238
Iron No.	TC	Si	Mn	Cr	Ac	Ar	Critical Temp., °F.	Expansion.	Ac-Ar °F.	
					Beginning, °F.	End, °F.	Beginning, °F.	End, °F.		
44	2.00	2.28	3.27	0.32	1427	1472	1148-932	212	1787	0.4 x 10^-3
45	2.27	4.52	2.35	0.47	1544	1706	1310-1040	527	1328	2.8 x 10^-3
46	2.23	3.78	2.60	0.48	1517	1562	1220-1112	347	1427	1.6 x 10^-3
Iron No.	TC	Si	Mn	Ac	End, °F.	Ar ₁ , °F.	Ar ₂ , °F.	Critical Temp., °F.	Expansion.	Ac-Ar °F.
				Beginning, °F.						
37	3.42	2.81	2.39		1472	1535	...	374	1472-1832	0.3-1.2 x 10^-3 1193
38	3.32	2.00	2.24		1436	1499	...	338		1193
39	3.22	2.77	2.01		1499	1553	1184	401		...
40	3.15	3.07	1.77		1526	1562	1202-1067	392		...
41	3.21	3.05	1.67		1508	1562	1202-1076	392		...
42	2.96	2.88	1.43		1490	1553	1220-986
43	2.90	2.73	1.29		1490	1553	1202-977

38. The iron containing 4.7 per cent manganese contained also some spots of cementite, but we cannot assert that all the manganese that was added to the ladle was completely dissolved (Figs. 8 and 9).

39. In Table 16 are given the thermal properties of a number of irons as determined with the Chevenard dilatometer. The position of the points *Ar* as indicated is that obtained after cooling in still air, and all the temperatures are given in Fahrenheit.

40. The expansion indicated is that produced by two cycles of heating, starting from the ambient temperature up to 1652 degs. Fahr. (900 degs. Cent.) in the first cycle and up to 1832 degs. Fahr. (1000 degs. Cent.) for the second, the heating and cooling of the furnace taking place at a rate which can be deduced from Fig. 10.

41. The figures given above show that the influence of manganese on the hardening by quenching of irons is small up to 1.54 per cent, but that as soon as the manganese content reaches 2 per cent the irons are air hardening for the rates of cooling adopted

during the tests. This characteristic, which can be used for the high-duty irons numbered 27 to 31, cannot be used for the irons numbered 37 to 43, the structure of which is in large grains (Fig. 6) and in which the graphite occurs in thick lamellae (Fig. 7). The mechanical properties of these last-named irons are not very good.

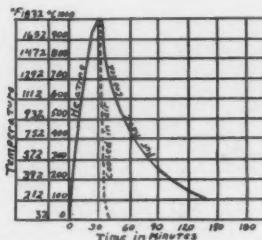


FIG. 10—HEATING AND COOLING RATES. HEATING: MEAN RATE BETWEEN 400 DEGS. CENT. (752 DEGS. FAHR.) AND 1000 DEGS. CENT. (1832 DEGS. FAHR.), 26 DEGREES PER MINUTE. COOLING: STILL AIR, MEAN RATE BETWEEN 1000 DEGS. CENT. AND 400 DEGS. CENT., 4.5 DEGREES PER MINUTE; IN FURNACE, MEAN RATE BETWEEN 1000 DEGS. CENT. AND 400 DEGS. CENT. 18 DEGREES PER MINUTE.

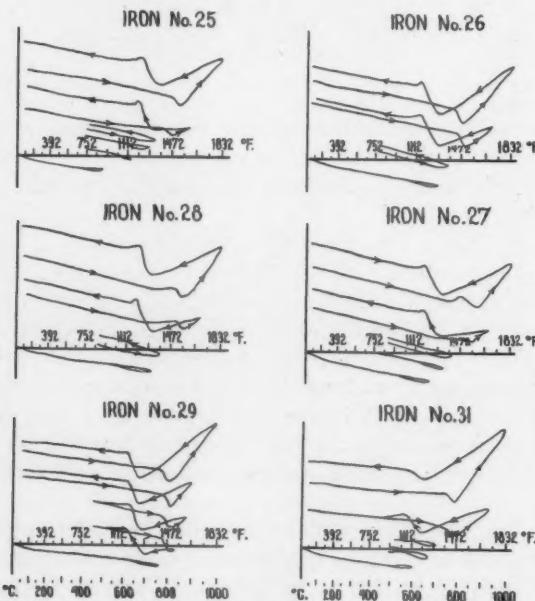


FIG. 11—STABILITY OF LOW CARBON AND HIGH MANGANESE IRONS.

42. One also can observe that the critical temperature, and therefore the stability of the irons, increases with the manganese content; so that the expansion for a range of heating, even reaching high temperatures, is low in the group of irons numbered 27 to 31, which factor is favorable for quenching treatment.

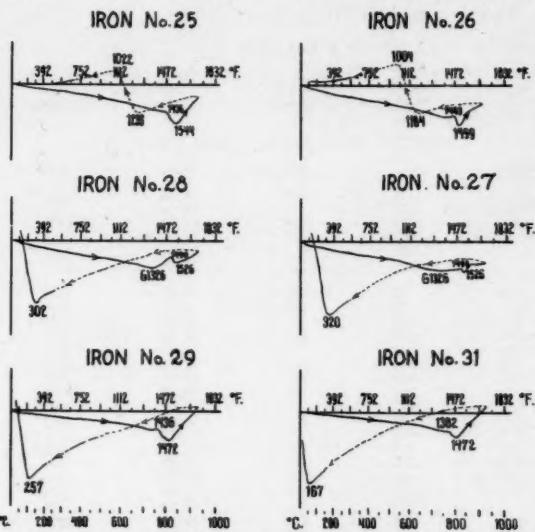


FIG. 12—APTITUDE FOR QUENCHING, OF LOW CARBON AND HIGH MANGANESE IRONS.

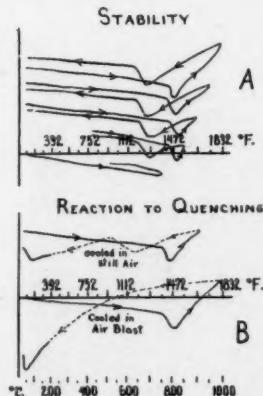


FIG. 13—MANGANESE-CHROMIUM IRONS. A: STABILITY. B: APTITUDE FOR QUENCHING.

EFFECT OF QUENCHING AND DRAWING ON SMALL TEST PIECES.

43. Following is a discussion of the mechanical characteristics obtained after quenching and drawing on pieces tested in the laboratory. Test pieces, as treated, had the final dimensions indicated earlier in this paper. The hardness tests were made on the test

Table 17

Treatment.	Drawing Temp., °F.	Brinell Hardness No.	Shear Strength, lbs. per sq. in.	Transverse Strength, lbs.	Deflection, mm.
IRON No. 29—Ac 1436-1472 DEGS. FAHR.					
Quenched in still air.....	600	600	76,788	1629	0.15
Quenched in blast at 1607 degs. Fahr., then drawn and cooled in still air	59 212 1112 1292 1436 1652	600 418 277 248 ...	62,560 82,476 85,320 51,192 63,990 74,944	1588 2139 2315 2205	0.11 0.16 0.17 0.19
Quenched in boiling water at 1607 degs. Fahr. then drawn and cooled in still air.	59 572 752 932 1112 1292 1436 1652	580 600 495 495 405 302 286 614	71,811 96,696 91,008 88,164 74,944 65,412 63,990 79,632	2117 2425 2205 2271 2007 2139 2646 1642	0.14 0.15 0.15 0.17 0.13 0.16 0.21 0.14
Quenched in oil at 1607 degs. Fahr., then drawn and cooled in still air.	59 212 572 752 932 1112 1292 1436 1652	580 340 690 495 444 418 241 241 600	78,176 92,430 96,696 96,696 85,320 74,944 65,412 59,724 99,540 to 85,320	2315 2425 2205 2271 2074 2151 2646 1818	0.15 0.15 0.15 0.17 0.13 0.16 0.21 0.14
IRON No. 27—Ac 1499-1526 DEGS. FAHR.					
Quenched in still air at 1607 degs. Fahr., then drawn and cooled in still air.	59 212 572 932 1292 1472 1652	340 332 332 321 277 255 311	65,412 65,412 71,811 66,123 62,560 59,724 71,811	2315 1874 2425 2315 2315 2425 2205	0.19 0.14 0.18 0.15 0.15 0.20 0.18
Quenched in air blast at 1607 degs. Fahr., then drawn and cooled in still air.	59 212 572 932 1292 1472 1652	500 332 495 418 277 260 311	74,944 78,210 93,852 85,320 65,412 62,560 71,811	992 882 2646 2205 2151 2205 2139	0.08 0.07 0.18 0.16 0.18 0.20 0.17
Quenched in boiling water at 1607 degs. Fahr., then drawn and cooled in still air.	59 212 572 932 1292 1472 1652	332 495 418 286 277 311	66,834 76,788 102,384 88,164 62,560 62,560 79,632	948 904 2425 2151 2205 2602 2096	0.07 0.06 0.17 0.14 0.15 0.21 0.18
Quenched in oil at 1607 degs. Fahr., then drawn and cooled in still air	59 212 572 932 1292 1472 1652	332 495 387 248 255 311	82,476 99,540 85,320 66,123 63,990 82,476	1433 2911 2249 2205 2448 2227	0.12 0.20 0.15 0.18 0.21 0.18

pieces used for the shearing test; the quenching temperature was 1607 degs. Fahr. (876 degs. Cent.) in every case, and the test pieces were maintained at this temperature and also at the drawing temperature for 15 minutes. (See Table 17.)

44. A comparison of iron No. 29 with irons Nos. 28 and 27 shows that the first iron gives higher hardness figures after quenching in still air as well as in the other cases, this being due, first, to the higher manganese content, which is 2.50 per cent instead of 2.10 per cent, and, second, to the lower graphite content. As to the

Table 17 (Continued)

Treatment.	Drawing Temp., °F.	Brinell Hardness No.	Shear Strength, lbs. per sq. in.	Transverse Strength, lbs.	Deflection, mm.
IRON No. 28—Ac 1490-1526 DEGS. FAHR.					
Quenched in still air at 1607 degs. Fahr., then drawn and cooled in still air.	59	321	78,176	2095	0.16
	212	311	74,944	1896	0.18
	572	293	90,297	2139	0.21
	932	286	74,944	2051	0.16
	1292	260	65,412	2051	0.15
	1472	241	62,990	2646	0.24
	1652	293	71,811	2139	0.18
Quenched in air blast at 1607 Degr. Fahr.	59	...	71,811	915	0.10
	212	...	79,632	948	0.09
	572	495	109,494	2426	0.21
	932	387	86,742	1896	0.11
	1292	286	65,412	2183	0.15
	1472	248	62,990	2603	0.28
	1652	293	72,522	1874	0.16
Quenched in boiling water.	59	...	82,598	904	0.08
	572	477	113,760	2425	0.23
	932	387	85,320	2117	0.20
	1292	269	65,412	2095	0.18
	1472	248	62,990	2470	0.21
	1652	293	72,522	1786	0.20
	59	...	72,522	962	0.10
Quenched in oil.	212	...	74,944	1103	0.11
	572	532	113,760	2646	0.20
	932	418	85,320	2095	0.14
	1292	277	65,412	2271	0.18
	1472	248	62,990	2381	0.22
	1652	293	73,233	2051	0.18
	59	...	72,522	962	0.10
IRON No. 45—Ac 1544-1706 DEGS. FAHR.					
Quenched in air blast, then drawn and cooled in still air.	59	430	68,687	1036	0.10
	302	...	83,182	2051	0.15
	572	...	91,008	2205	0.15
	752	375	78,770	2403	0.20
	932	364	61,430	1764	0.14
	1112	293	57,440	1742	0.12
	1292	255	50,708	1676	0.11
	1472	262	62,074	1786	0.17
	1652	311	65,696	1367	0.11
	59	...	63,990	1345	0.10
IRON No. 46—Ac 1517-1562 DEGS. FAHR.					
Quenched in air blast, then drawn and cooled in still air.	59	...	103,806	2182	0.12
	302	...	108,925	2249	0.13
	572	405	100,678	2756	0.15
	752	418	85,888	2271	0.14
	932	418	71,811	2139	0.15
	1112	321	62,990	2073	0.15
	1292	277	84,036	2448	0.16
	1517	402	71,811	1962	0.14
	1652	477

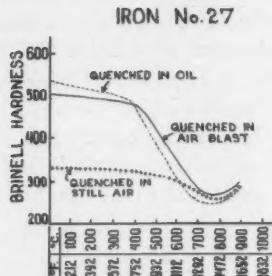


FIG. 14—HARDNESS CURVES AFTER QUENCHING AND DRAWING, FOR IRON NO. 27. COMPARISON OF QUENCHING IN STILL AIR, IN AIR BLAST AND IN OIL.

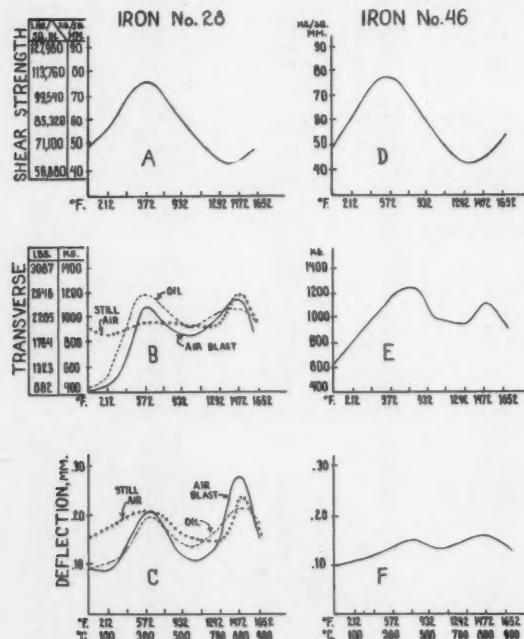


FIG. 15—EFFECT OF QUENCHING AND DRAWING ON THE MECHANICAL PROPERTIES OF IRONS 28 AND 46.

curves representing the results of mechanical tests in relation to drawing temperatures, it will be noted that these curves show consistently the same shape.

Hardness.

45. The hardness decreases very slowly for drawing temperatures up to 752 degs. Fahr. (400 degs. Cent.); beyond that temperature, the hardness drops quite rapidly and then increases again when the drawing temperature exceeds the transformation point.

Shearing Strength.

46. The shearing strength increases in every case for a drawing temperature of 572 degs. Fahr. (300 degs. Cent.); it then decreases progressively up to the drawing temperatures which are near to the transformation point, after which there is an increase due to quenching, the draws being followed by cooling in still air (Fig. 15, *A* and *D*).

Transverse Strength.

47. The transverse strength increases rapidly between 32 and 572 degs. Fahr. (0 and 300 degs. Cent.), but in many cases there is a slight drop for a draw at 212 degs. Fahr. (100 degs. Cent.), particularly in the case of iron No. 27. Beyond 572 degs. Fahr. the transverse strength decreases progressively up to 1292 degs. Fahr. (700 degs. Cent.) and then increases again up to 1472 degs. Fahr. (800 degs. Cent.), which is the temperature near to the transformation point. The transverse strength then drops again when the drawing temperatures exceed the transformation point and when the metal is affected by quenching in still air during cooling (Fig. 15, *B* and *E*).

Deflection.

48. It is remarkable that the deflection curves have the same shape as the transverse strength curves, with the result that the irons which offer the greatest resistance to deformation are those which are the most liable to deformation before breaking. It should be noted that the deflections have a tendency to be highest for those drawing temperatures which are closest to the critical temperature. This is easily conceived if these results are compared with those that would be obtained in the resilience test of heat-treated steels (Fig. 15, *C* and *F*).

49. In fact, one notes two particularly remarkable tempera-

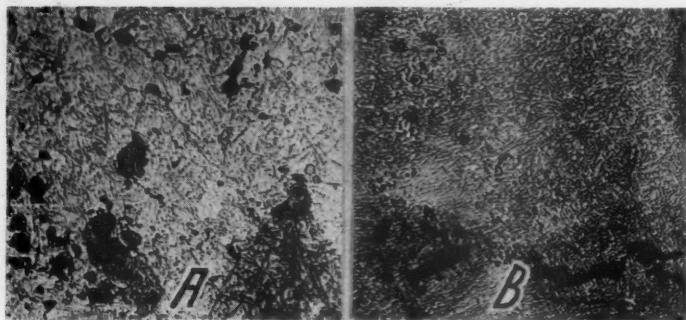


FIG. 16—PHOTOMICROGRAPHS OF IRON NO. 29. ETCHED WITH NITRIC ACID, X500. A: FRACTURE AFTER QUENCH AT 875 DEGS. CENT. (1607 DEGS. FAHR.), COOLED IN STILL AIR, SHOWING GRAPHITE, MARTENSITE AND A LITTLE TROOSTITE. B: FRACTURE AFTER QUENCH IN BOILING WATER AND DRAWN AT 300 DEGS. CENT. (572 DEGS. FAHR.) THE SHADING AROUND THE GRAPHITE IS DUE TO OXIDATION RESULTING FROM THE ETCHING.

tures are apparent, 572 and 1472 degs. Fahr. (300 and 800 degs. Cent.) The point corresponding to 1472 degs. Fahr. could be foreseen by analogy with steels; the point corresponding to 572 degs. Fahr. cannot be so easily explained, at least so far as shearing strength is concerned. Probably it coincides with the disappearance of the martensite, which would explain the greater liability to deformation of the metal and, consequently, the higher trans-

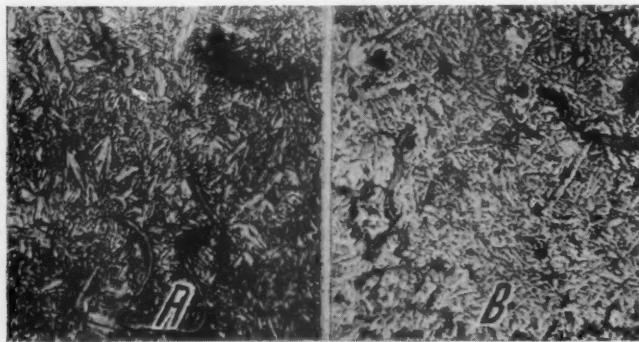


FIG. 17—PHOTOMICROGRAPHS OF IRON NO. 27. QUENCHED IN AIR BLAST AT 875 DEGS. CENT. (1607 DEGS. FAHR.) ETCHED WITH NITRIC ACID, X400. A: DRAWN AT 300 DEGS. CENT. (572 DEGS. FAHR.), STRUCTURE SHOWING GRAPHITE, TROOSTITE AND REMAINS OF MARTENSITIC NEEDLES. B: DRAWN AT 800 DEGS. CENT. (1652 DEGS. FAHR.), STRUCTURE SHOWING GRAPHITE AND SORBIT.

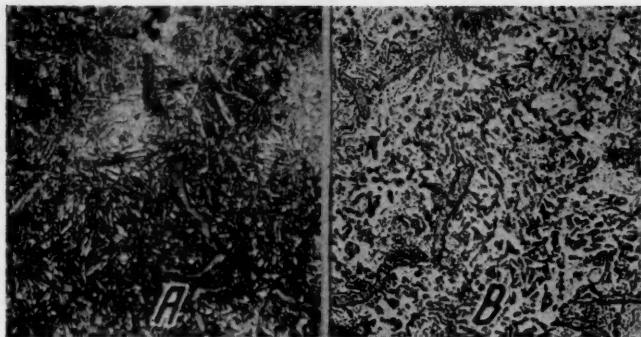


FIG. 18—PHOTOMICROGRAPHS OF IRON NO. 27. QUENCHED IN OIL AT 875 DEGS. CENT. (1607 DEGS. FAHR.) ETCHED WITH NITRIC ACID, X400. A: DRAWN AT 300 DEGS. CENT. (572 DEGS. FAHR.), STRUCTURE SHOWING GRAPHITE, TROOSTITE AND REMAINS OF MARTENSITIC NEEDLES. B: DRAWN AT 800 DEGS. CENT. (1652 DEGS. FAHR.) STRUCTURE SHOWS GRAPHITE, GLOBULAR PEARLITE AND FERRITE.

verse strength and the greater deflections obtained after drawing at 572 degs. Fahr. (300 degs. Cent.)

50. However, in the case of shearing strength we can only conceive the low figures obtained between 32 and 572 degs. Fahr. (0 and 300 degs. Cent.) by the considerable brittleness of the martensite, which would cause premature ruptures during the shearing test. The real curve obtained by extrapolation should then have a shape identical to that of heat-treated steels, and the maximum obtained at 572 degs. Fahr. would then indicate the disappearance of the brittleness. The draws not having been affected between 212 and 572 degs. Fahr. (100 and 300 degs. Cent.), it is probable that this maximum would be slightly displaced toward

Table 18
BRINELL HARDNESS

Iron No.	Test Pieces	As Cast	Quenched at 1697° F.		
			Drawn at 1292° F.	Cooled in Still Air	Cooled in Air Blast
20	0.866 x 0.709 x 0.62 in. (22 x 18 x 16 mm.)	340	277	532	532
27	1.57 in. (40 mm.) cubes	321	255	351	532
28	1.57 in. (40 mm.) cubes	311	260	351	532
29	1.57 in. (40 mm.) cubes	332	286	364	532
			Drawn at 1382° F.	Cooled in Still Air	Cooled in Air Blast
30	Gear.....	332	248	340	512
30	Steam Engine Liner ..	293	245	...	512
					Cooled in Oil

the left, and should coincide with the total disappearance of the martensite toward 392 degs. Fahr. (200 degs. Cent.) for prolonged periods of drawing.

Practical Tests

51. The results of laboratory tests given in this paper show the great interest that can be presented by irons with low carbon and with a manganese content of 2 per cent or above. These results have been confirmed by some practical tests, or tests made under conditions nearly practical. Table 18 shows the results of hardness

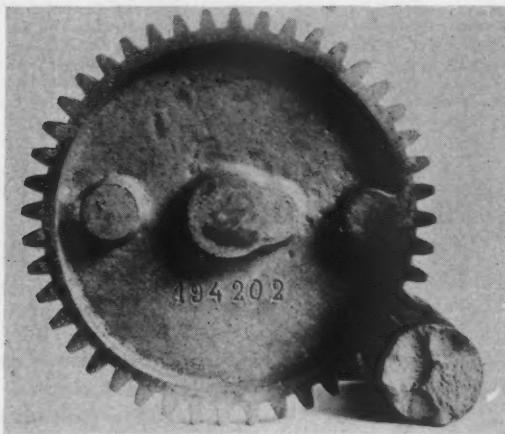


FIG. 19—IRON NO. 30, CAST INTO A GEAR, SHOWING FRACTURE OF GATE. HEAT TREATMENT AND BRINELL HARDNESSES: AS CAST, BRINELL 332; DRAWN 3 HOURS AT 750 DEGS. CENT. (1382 DEGS. FAHR.), BRINELL 248; QUENCHED IN STILL AIR AT 950 DEGS. CENT. (1697 DEGS. FAHR.), BRINELL 240; QUENCHED IN AIR BLAST AT 925 DEGS. CENT., BRINELL 512.

tests obtained on pieces larger than those indicated above, and on a gear made from iron No. 30.

Practical Conclusion

52. The results of quenching in an air blast show this to be the method to be adopted in practice for pieces of medium size. On the other hand, the laboratory tests on small test pieces show that it is advisable, when great hardness with sufficiently good machinability is required, to draw after quenching at a tempera-

Table 19
Machinery Tests

Tool Material	Machining Speed		Remarks
	Feet	Meters	
IRON No. 29—BRINELL HARDNESS OF TEST PIECE AS CAST, 335.			
High-speed steel R. S. V. from the Saar..	39.36	12.0	Tool broke.
High-speed steel R. S. V. from the Saar..	29.52	9.0	Tool broke.
High-speed steel R. S. V. from the Saar..	17.04	5.5	Tool lasts 5 minutes.
Tungsten carbide.....	65.60	20.0	Works normally.
Tungsten carbide.....	104.69	32.0	The tool stood up best; test stopped on account of vibration.
IRON No. 30—BRINELL HARDNESS OF TEST PIECES AS CAST, 293 TO 321.			
High-speed steel R. S. V. from the Saar..	21.32	6.5	The tool cut but was rapidly worn.
Tungsten carbide.....	75.44	23.0	Works normally; could do better.
High-speed steel with 8 per cent Cobalt..	59.04	18.0	Works normally; tool wears rapidly.
IRON No. 30—DRAWN AT 1382 DEGS. FAHR. AND COOLED IN FURNACE, BRINELL HARDNESS 241			
High-speed steel R. S. V.....	39.36	12.0	Normal.
High-speed steel R. S. V.....	52.48	16.0	The tool cuts but wears very rapidly.
Tungsten carbide.....	140.88	46.0	Stands normally not withstanding vibration.
Tungsten carbide.....	246.00	75.0	Stands normally; this does not permit greater speed.
High-speed steel with 8 per cent Cobalt..	65.60	20.0	Stands normally.
High-speed steel with 8 per cent Cobalt..	85.28	26.0	Tool cuts but wears rapidly.

ture at least equal to 572 degs. Fahr. (300 degs. Cent.) This corresponds to what has been found by Guillet, Galibour and Ballay on nickel cast irons with a high content of carbon.

MACHINABILITY

53. All the machinability tests were made on a small lathe with a feed of 0.039 in. (1 mm.) and a cut of 0.068 in. (1.75 mm.) on test pieces having a length of 15.75 in. (400 mm.) and a diameter of 1.3 in. (45 mm.) The length and the small diameter of the machine bars were a considerable drawback, owing to the vibrations occurring at great speeds. (See Table 19.)

54. When drawn, this iron (No. 30) can be machined like ordinary cast iron; it will be seen, therefore, that the great hardness in the natural state, of irons with a high manganese content of which the analysis is judiciously chosen, cannot be considered as an obstacle to their machinability. Furthermore, it has been possible to machine on a vertical lathe some steam engine cylinder linings made with the iron No. 30 and having an exterior diameter of 11.5 in. (290 mm.) and a thickness of 1.06 in. (27 mm.) when



FIG. 20—IRON NO. 30, CAST AS STEAM-ENGINE CYLINDER LINER. THIS CASTING WHICH WOULD HAVE TAKEN A MARTENSITIC STRUCTURE BY AIR QUENCHING WAS QUENCHED IN OIL AS AN EXPERIMENT. BRINELL HARDNESS: AS CAST, 293; AFTER DRAWING, 248; AFTER AIR QUENCH, 512.

Table 20

MACHINING OF CYLINDER LININGS OF IRON NO. 31-C CONTAINING 4.7 PER CENT MANGANESE WITH CASTINGS HEAT TREATED BY DRAWING.
BRINELL HARDNESS, 310

Tool Material	Tool Speeds and Feeds						Remarks
	Speeds		Feeds				
	Meters	Feet	In.	mm.	In.	mm.	
High-speed steel R. S. V....	9	29.52	0.049	1.25	0.157	4	Tool did not bite.
High-speed steel R. S. V....	5	16.40	0.049	1.25	0.157	4	Normal cutting with the tool standing up well.
Tungsten carbide.....	14	45.92	0.032	0.83	0.079	2	Normal cutting.
Tungsten carbide.....	32	104.96	0.032	0.83	0.079	2	Edge of the tool became red hot but cut normally.

cast (with Brinell hardness of 286 to 293—see Fig. 20), under conditions similar to those obtaining in ordinary circumstances. (See Table 20.)

SUMMARY

55. Summarizing the above study, manganese definitely favors the martensitic quenching as soon as it exceeds 2 per cent, to such an extent that medium-size pieces with a thickness of 1.06 in. (27 mm.) could take a martensitic structure simply by cooling in still air.

56. On the other hand, the low carbon content has improved the mechanical properties and has considerably reduced the danger of microscopic cracks or fissures due to rapid cooling, and this notwithstanding the presence of a high manganese content. These proportions of low total carbon and high manganese have contributed to the increased stability of irons subjected to heat, which property is to be looked for in irons to be subjected to heat treating.

57. Very low contents in total carbon should be avoided in current practice, owing to an increase in difficulties of casting, to the rapidity with which the metal passes from the liquid to the pasty state, and to its aptitude to piping. A content of about 2.50 per cent can be adopted when melting iron in the cupola, while in the electric furnace, which enables the iron to be superheated, one should obtain interesting mechanical properties with higher carbon, which can be around 2.80 per cent.

58. Finally, if one considers the investigations of Guillet, Galibourg and Ballay, one can foresee interesting combinations that can be used in practice in the preparation of nickel-manganese irons with low total carbon content and high manganese content. Similarly, one could expect to see the nickel neutralize the action of the manganese, which in any case is slight, on the chilling in the mold, while their action on the aptitude to martensitic quenching would be added to each other, with the result of giving irons which have a particular tendency for quenching. Furthermore, the low carbon contents, or else super-heating, would contribute to the decrease of the danger of microscopic cracks or fissures, a danger which we have not, as a matter of fact, come across during our investigations on pieces having a simple shape, the results of the mechanical tests not having indicated any particular depreciation in the qualities of the iron, but rather a noticeable improvement.

ACKNOWLEDGMENT

59. The author wishes to acknowledge the considerable help which he has received from the use of the differential dilatometer by Pierre Chévenard, whose apparatus is invaluable in the scientific industrial researches effected in the metallurgical laboratories of France. Acknowledgment is also made of the previous work of Guillet, Galibourg and Ballay.

Air-Furnace Cast Iron

BY DUNCAN P. FORBES,* ROCKFORD, ILL.

Abstract

The design and production methods for making cast irons in the air furnace are first described. The importance of skillful operations is discussed and points to be observed are listed. Duplexing, with the air furnace used for refining and superheating, is discussed extensively. The production of high-test iron is described, stress being laid on carbon control. Compositions, structures and properties obtained in such irons are given, and the use of alloys reviewed.

1. So much interest has been shown in improving quality of cast iron in the last few years that it is fitting to consider and describe what is probably the oldest method of making high-strength cast iron in this country. The air furnace, or reverberatory furnace, has been in use a great many years for the melting of several different metals, and one of the earliest applications was in the gray iron field to produce castings of high strength. Without going into the history of the air furnace, it is interesting to consider the furnace in its present form and also the nature of the gray iron which is now being produced in it in this country.

AIR FURNACE DESIGN

2. Air furnaces fall into three general classifications, depending on the type of fuel used in the melting operation. The type of furnace which has been in use the greatest number of years is that using hand-fired coal for fuel. In general design, a furnace consists of a more or less rectangular hearth with a fuel combustion chamber at one end and a stack at the other end.

3. When hand-fired coal is used, the combustion chamber takes the form of a fire-box fitted with grate bars and equipped with a blast pipe below the grate bars for forcing air through the fuel bed, which is from 12 to 18 inches in thickness. The coal used is usually of high B.T.U. content, with a high percentage of

*Guntine Foundries Corp.

NOTE: This paper was presented and discussed before one of the gray iron sessions at the 1933 Convention of American Foundrymen's Association.

volatile matter, and the resulting flame is very rich in unburned gas and carbon monoxide.

4. The flame passes over a low bridge wall where a quantity of secondary air is blown in through tuyeres in a wind bung located over the bridge wall. This secondary air completes the combustion of the raw, gaseous flame and directs the resulting high-temperature flame down upon the charge to be melted and upon the molten bath which results therefrom.

5. The flame travels through the furnace rather rapidly, heating the charge by radiation as well as by conduction, and passes out to the stack over a low bridge wall at the rear end of the furnace.

6. The heat balances of air furnaces and the melting efficiency have been quite completely covered in previous papers dealing with this subject. For our purposes it is only necessary to describe the types of furnaces as an introduction to the consideration of air-furnace cast iron, which is to follow. It is sufficient, in passing, to say that from a fuel efficiency standpoint the air furnace is extremely wasteful.

7. Within the last twenty years the use of pulverized coal and fuel oil have become almost universal in air furnaces in the malleable industry, with the greatest popularity being shown the furnaces using pulverized coal. In construction, the pulverized-coal furnace differs from the hand-fired coal furnace in that the fire box is eliminated and in its place one or more pulverized coal burners are located in the end wall of the furnace, blowing the pulverized coal flame downward at a slight angle upon the hearth. (See Fig. 1). No secondary air is necessary to complete the combustion.

8. The oil-fired furnace is similar to the pulverized coal furnace, except that a conical combustion chamber about six or eight feet in length is usually added at the firing end of the furnace to confine the oil flame and to assist the combustion of the oil spray and the air blast. Most of the oil-fired furnaces in use at the present time provide means of admitting secondary air to the flame at the end of the combustion chamber through tuyeres in a wind bung, the amount of air admitted at this point in relation to the amount that is blown in with the oil spray varying widely among different operators. Successful installations are in operation where the secondary air varies from one-third to two-thirds of the total amount of air used.

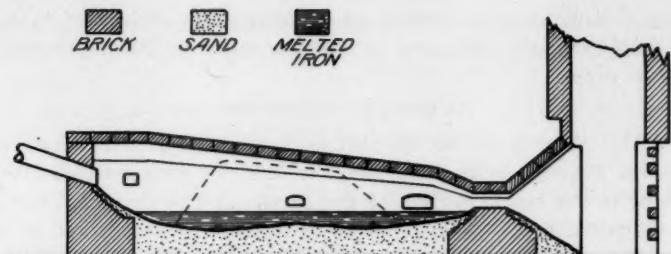


FIG. 1—PULVERIZED-COAL-FIRED AIR FURNACE SHOWN IN LONGITUDINAL SECTION WITH STEELWORK OMITTED. DOTTED LINE SHOWS SPACE OCCUPIED BY COLD CHARGE.

Table 1
TYPICAL ANALYSIS OF 15-TON CHARGE FOR HAND-FIRED AIR FURNACE

Material.	Lbs.	Charge.	Si	C	Mn	S	P
Remelt	15,000	50	2.00	2.90	0.45	0.060	0.160
Pig iron	7,500	25	1.50	4.00	0.80	0.035	0.180
Pig iron	7,500	25	4.40	4.00	0.94	0.022	0.150
Total Charge	30,000	100	2.48	3.40	0.76	0.045	0.163
Iron tapped	28,500	95	2.03	2.90	0.51	0.063	0.165
Per cent Loss of Elements.....		18.1	14.7	32.9	40.0*	1.2*	

*Gain.

9. Without discussing the merits of one type of furnace over another, it can be stated that the oil-fired and pulverized-coal-fired furnaces are more popular than hand-fired coal furnaces because of their rapid rate of melting and the fact that a low-priced charge can be used to obtain a given ultimate analysis. Fuel economy is also an important factor.

AIR FURNACE PRODUCTS

10. In America, air furnaces are used principally for the production of white-iron castings which are later to be annealed into malleable iron. For many years, however, high-grade gray iron has also been produced by the air furnace process, because this process offered the simplest means of obtaining low-carbon metal. In recent years, with the developments in metallurgy due to the use of alloys and electric-furnace melting, air-furnace cast iron has assumed greater importance. This metal is called "gun iron," due to its use of ordnance in the middle of the last cen-

tury. "Gun iron" is defined by Webster as "cast iron of extraordinary strength, prepared by melting in an air furnace instead of the cupola."

PRODUCTION METHODS

11. Melting in the air furnace is accomplished by first removing a section of the roof bungs or firebrick arches, which gives access to the hearth section of the furnace. The charge of metal is deposited on the sand or brick bottom of the hearth and is so distributed that melting will be uniform across the width of the hearth. The charge usually consists of remelt from previous heats, purchased scrap, and pig iron. Table 1 shows the typical charge. After the charging operation is complete, the bungs are replaced and sealed up with fireclay. The fire is then lighted in the fire box, or the pulverized coal or oil burners are lighted.

12. Melting technique varies widely through the plants using air furnaces, some operators preferring to melt as rapidly as possible with a sharp, hot flame. Some prefer to use a soft flame and take a longer time to accomplish the melting-down stage of the operation. Others prefer to soak the charge at a moderate temperature, in order to preheat it to the bottom of the hearth before any great amount of melting occurs. After the charge has been fairly well preheated, melting is then conducted as rapidly as possible.

13. When the charge is in the all-molten condition, it is almost universal practice to fire as hard as possible to give the metal enough superheat for pouring in the shortest possible time. This will require from one to two hours, and the molten metal is

Table 2
TIME SCHEDULE FOR 15-TON HEAT, HAND FIRED

Time.	Operation.
0:00.....	Light fire.
0:45.....	Turn on bottom blast and partial top blast.
1:45.....	Top blast increased to full.
4:45.....	Stir and skim slag.
5:45.....	Stir and take sample for analysis.
6:15.....	Stir and pour fracture specimen.
6:45.....	Stir and tap furnace.
6:50.....	Reduce top blast.
7:30.....	Furnace empty.

usually stirred and skimmed several times during this period. Table 2 gives a time schedule for a 15-ton heat in a hand-fired furnace.

14. Some operators prefer to leave the slag on the metal until just before tapping, thinking thereby to protect the metal from the oxidizing action of the flame. However, in most plants the metal is skimmed one or more times during the heating-up period, on the assumption that the presence of slag on the surface of the metal would delay the heating-up period to such an extent that there would be no gain, as far as avoidance of oxidation is concerned.

15. When the metal is in the molten condition and has become sufficiently superheated, it is customary in all plants to pour a specimen or bar which can be broken so that the melter can examine the fracture. In some plants chemical analysis is also made from specimens taken at this time. When the tests have been completed, if the metal is of a correct composition and temperature, pouring can begin immediately. If the metal is not of correct analysis and fracture, the melter can make the necessary additions, of steel, alloys or petroleum coke to obtain the exact composition desired.

16. In the air furnace the carbon, silicon and manganese are being oxidized at all times, and frequently, if tests indicate a surplus of these elements, an additional ten or fifteen minutes of firing will bring the metal to the desired analysis. When melting with pulverized coal, some of the more coarse coal particles often will fall into the bath and be absorbed, resulting in a slight pick-up of carbon which will counteract the loss due to oxidation. In general, this carbon pick-up is looked upon as dangerous, although under most circumstances no harm would result from a moderate amount of this carbon absorption.

IMPORTANCE OF SKILLFUL OPERATION

17. The above description of the melting operation is, naturally, very brief, and it is not too much to say that the experience and technique of the melter will have a great bearing on the uniformity and quality of the resulting metal.

18. The important points about successful air-furnace melting which must be observed are:

- (a) Design and proportions of the inside of the furnace.

- (b) Selection of materials and melting stock.
- (c) Scrupulous adherence to a definite program of operation.
- (d) Skill and experience in testing the metal before tapping, and judgment in applying correctives, if needed.
- (e) Attention to resulting physical and chemical properties of metal from each heat, so that minor variations can be corrected in subsequent heats before serious divergence from standard practice occurs.

DUPLEXING WITH THE AIR FURNACE

19. A variation of operation of the air furnace occurs when the air furnace is used only for refining and superheating the metal instead of as a means for melting. Great progress has been made in the United States in duplexing gray iron and malleable iron, where the metal is melted in a cupola and refined in an electric furnace or an air furnace.

20. To understand the reasons for duplexing, it is necessary to consider briefly the capabilities of a cupola. The cupola provides the least expensive known means of melting iron. The thermal efficiency is excellent, and the furnace, if the investment involved is considered, has a very high melting rate. For melting small heats the cupola is not particularly efficient, but where melting can be continued for many hours the cost is very low.

21. A cupola can be operated either with a continuous flow of metal at the spout or by intermittent tapping of individual charges. In either case, the analysis of the metal flowing from the spout is not uniform, due to the fact that the charge does not melt uniformly and that the carbon and silicon may vary over a considerable range. This variation is minimized if individual charges are tapped intermittently, because the metal which accumulates on the bed of the cupola has an opportunity to become somewhat uniform.

22. If we are to consider only high-test iron, which is relatively low in carbon, there is a decided advantage in having the cupola deliver metal continuously, since a minimum time then is allowed after melting for carbon pick-up from the coke bed.

23. In cupola operation, metal passes from the solid to the molten condition in a remarkably short period of time. In the melting zone the small drops of metal which are melted from the

pig iron and scrap will trickle down through the bed of incandescent coke, reaching the bottom in a few seconds' time in a condition hot enough for pouring. Investigators have established the fact that metal of this kind does not seem to graphitize during solidification in the same manner as metal melted in an air furnace or in an electric furnace. Without attempting to explain the reason for this phenomenon, we can state that time and temperature probably are the controlling factors.

24. In the case of duplexing with an electric furnace, the metal is superheated sufficiently to eliminate any tendency toward coarse graphite formation and, if desired, a time element can also be provided by holding the metal for a period of time in the electric furnace before pouring. In the case of the air furnace, a very high temperature can also be obtained; but usually the melt heats up at a relatively slow rate and the air furnace provides a long enough period of time so that the same result can be accomplished without going to as high a temperature as is required in the electric furnace.

25. The reasons for duplexing, therefore, are to take advantage of the low melting cost of a cupola and at the same time provide the necessary degree of superheat, or the necessary period of time in the molten condition to accomplish the refinement of the graphitizing tendency of the metal. There is an additional function of the air furnace in that a reduction of the total carbon can be obtained by steel additions or by mild oxidation in case the cupola does not bring the metal down with a low enough carbon content. There is also the advantage that preliminary analyses and corrections can be made if the cupola operates at an analysis not desired.

26. Duplexing with an electric furnace is not within the scope of this paper. Therefore, I will attempt to outline only the procedure used in duplexing with an air furnace. The cupola can be placed adjacent to or some distance away from the air furnace, depending on the most convenient arrangement. The most logical arrangement is to have the cupola close enough to the air furnace so that the molten metal can be run by gravity through a long spout direct from one to the other.

27. It is also advantageous to have a receiver located between the cupola and the air furnace to hold an amount of metal equivalent to one cupola charge, so that the metal flowing in will mix with the metal already in the receiver and the metal flowing

out will be of substantially uniform composition. The metal can also be slagged at this point and a desulphurizing agent applied. Using approximately 6 lbs. of fused soda ash to the ton, the sulphur can be reduced from 0.125 to 0.090 per cent. Further desulphurization can be obtained using a greater amount of soda ash.

28. With such an arrangement the air furnace can be operated either for batch pouring or continuous pouring. If it is operated as a batch furnace it is usually desirable to permit the metal to remain under fire in the air furnace for approximately two hours. This permits ample time for preliminary analyses and the breaking of test samples before any metal is poured into castings. It also allows the metal to remain quiescent in a shallow bath so that slag and impurities may rise to the surface. Care should be taken to avoid excessive temperatures, as time and not temperature seems to produce the best results from air-furnace metal, particularly in the case of low-carbon gray iron.

29. If the air furnace is operated with continuous pouring, the body of metal in the furnace should approximate the amount that will be poured over a 2-hour period. This permits the metal to average two hours in the molten state; and while, theoretically, the incoming metal from the cupola may be adding a condition promoting the formation of coarse graphite, in actual operation no trouble is experienced from this cause.

30. In duplexing, the metal entering the air furnace must be high enough in carbon, silicon and manganese so that a 2-hour period of mild oxidation will bring the analysis to that ultimately desired. Despite this requirement, considerable leeway is possible in the selection of a charge for the cupola.

31. When producing high-test gray iron by duplexing, the charge will consist of remelt, plus a small amount of pig iron, some steel, and some purchased gray iron or malleable scrap. The operating characteristics of the cupola will, of course, determine the exact proportion to use. If desired, metal can be made entirely from steel scrap or bundled steel, making the proper analysis corrections by additions to the air furnace. Some experimentation is necessary to select the best charge to give the correct ultimate analysis, because no two cupolas and no two air furnaces operate in exactly the same manner, and raw material prices vary in different localities.

32. The economy of duplexing depends on the cost of materials, the size of the equipment, and the amount to be melted in a

day. It is doubtful if any saving would be made over cold melting in an air furnace unless the amount of metal to be used in a day exceeds thirty tons.

AIR FURNACE GRAY CAST IRON

33. Almost any type of cast iron can be made satisfactorily in an air furnace, but the air furnace is more commonly applied to the production of iron with less than 3.00 per cent carbon, because the cupola can produce this iron only with difficulty. For certain purposes, however, iron with carbon above 3.00 per cent is produced, and this iron can be made with great uniformity and purity. By using about 50 per cent pig and the balance remelt in the charge, sulphur can be held in the range 0.02 to 0.04 per cent and the resultant metal is close grained, with excellent physical properties.

34. For the production of true high-test iron it is customary to keep the carbon content at a point below 3.00 per cent. The low carbon is necessary so that a minimum of graphite will be present in the casting. Air-furnace iron, like cupola iron, requires that, for a given metal section, the silicon content should be increased if the carbon content is lowered. As a result, most high-test iron of less than 3.00 per cent carbon is produced with silicon of 2.00 per cent or more.

35. High physical properties in cast iron are favored by the following:

- (a) Low percentage of graphite.
- (b) Graphite of uniform size and distribution.
- (c) Graphite of a flake structure where each flake is separate from every other flake, permitting continuity of metallic matrix.
- (d) Absence of ferrite crystals in the microstructure.
- (e) Absence of cementite crystals.
- (f) Pearlitic or sorbitic structure in the matrix.
- (g) Presence of alloys of the type which add to the strength of steel.

36. High-test iron can be made with a carbon content as low as 2.00 per cent, although it is believed that none of this metal is produced commercially at the present time. Rarely is high-test iron made with less than 2.50 per cent.

37. The influence of carbon content on high-test iron is very great, considering only those irons which have a completely pearlitic matrix. It is found that metal with less than 2.60 per cent

carbon develops a dendritic arrangement of the graphite crystals, which becomes more pronounced as the carbon content is lowered. This results in a "woody" appearance of the fracture and an embrittlement which largely nullifies the advantages gained from a reduction in the total amount of graphite present in the castings.

38. In order to maintain a completely pearlitic structure, it is necessary to increase the silicon content of the metal as the carbon is lowered. Light-section castings will require an excessive amount of silicon, which will affect the heat-treating properties of the castings by reducing the percentage of combined carbon which can be retained in solution in austenite at temperatures above the critical.

Table 3
ANALYSIS AND PHYSICAL PROPERTIES OF TYPICAL AIR-FURNACE CAST IRON

Sample.	Si	C	Mn	Ni	Cr	Mo	Tranverse Strength,* lbs.	Deflection, in.	Brinell Hardness.
1	2.04	2.80	0.62	6100	0.19	196
2	2.20	2.54	0.54	6000	0.17	220
3	1.96	3.10	0.58	5600	0.18	187
4	2.04	2.80	0.62	0.42	6300	0.15	235
5	2.04	2.80	0.62	1.00	6300	0.16	217
6	2.04	2.80	0.62	0.80	6600	0.21	212

*Tranverse tests were conducted on a bar 1 1/4 inches in diameter, broken on 12-inch centers.

39. As the carbon is lowered, the hardness of the metal has a tendency to increase. For certain purposes this is very desirable, but it constitutes a handicap from the standpoint of machinability. If resistance to wear is the desired property of the metal, an investigation should be made to determine whether or not increased hardness brings about the desired result. Table 3 shows the analyses and physical properties of some typical air-furnace cast irons, and in Figs. 2, 3 and 4 are shown photomicrographs illustrating structures.

USE OF ALLOYS

40. The use of the air furnace permits very convenient use of alloys. The metal leaves the furnace hot enough to melt alloy additions quickly and permit time for thorough mixing in a ladle. Furthermore, the uniformity of analysis of air-furnace iron permits much more exact metallurgical practice than is the case with cupola metal. It must be remembered, however, that in applying

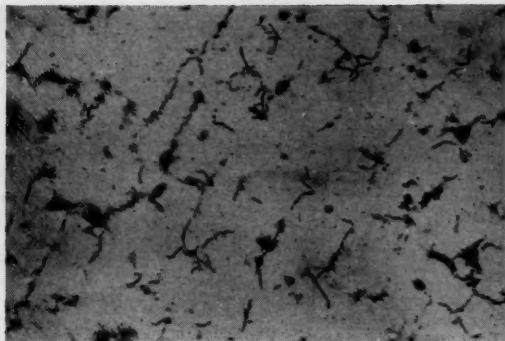


FIG. 2—PHOTOMICROGRAPH OF AIR-FURNACE CAST IRON, APPROXIMATELY 2.40 PER CENT CARBON. UNETCHED, X100.



FIG. 3—PHOTOMICROGRAPH OF AIR-FURNACE CAST IRON, APPROXIMATELY 2.80 PER CENT CARBON. UNETCHED, X100.

alloys to air-furnace iron, only the alloying properties are of benefit because the graphitizing tendency of the metal has already been taken care of by the melting process. Characteristics of some of the more common alloys are given below.

Molybdenum.

41. Molybdenum additions to air-furnace iron in amounts up to 0.50 per cent result in an increase of strength and deflection, with no substantial change in hardness. Percentages in excess of 0.50 continue to increase the strength, with a corresponding increase in hardness.



FIG. 4—PHOTOMICROGRAPH OF AIR-FURNACE CAST IRON, APPROXIMATELY 2.80 PER CENT CARBON. ETCHED WITH NITAL, X1000.

Nickel.

42. Nickel additions to air-furnace iron in small amounts up to 0.50 per cent produce very little change in the metal. Additions above this amount tend to increase the strength and hardness of the metal. The nickel does not produce any noticeable effect on the graphitizing tendencies.

Chromium.

43. Chromium additions are immediately effective in increasing the hardness and ultimate strength. If percentages of more than 0.50 are to be used, the metal will chill badly in thin sections and the microstructure will contain extensive deposits of undecomposed cementite. This can be avoided if the metal is produced with a higher silicon content, or if nickel is added with the chromium to provide an increased graphitizing tendency in the metal.

Manganese.

44. Manganese additions in moderate amounts have very little effect on air-furnace iron, except in heavy-section castings. In this case, higher manganese tends to prevent formation of ferrite crystals during the slow cooling below the critical temperature.

Hardness of Air Furnace, Cupola and Electric Iron

45. The question is often asked, wherein does air-furnace iron differ from cupola iron and electric-furnace iron. This is a

difficult question to answer, although there is evidence that a considerable difference exists. Some very high strength irons have been made in both the cupola and electric furnace without the use of alloys. In general, however, these irons are inclined to be quite hard, which is not the case with properly made air-furnace iron.

46. Air-furnace iron of 50,000 lbs. tensile strength can be produced with a Brinell hardness of 200 or less, whereas cupola and electric-furnace iron with a tensile strength of 50,000 lbs. usually has a Brinell hardness of 220 or more. The reason for this difference is not readily apparent.

47. Much of the success of air-furnace iron can be attributed to the fact that the few foundries in this country which produce it have available skilled metallurgists and laboratory equipment to insure intelligent manufacture and uniformity.

SUMMARY

48. The air furnace is the oldest means for the production of high-test cast iron.

49. Under certain conditions, economies of manufacture can be brought about by duplexing from the cupola to the air furnace.

50. Air-furnace iron differs from cupola iron in strength and hardness and differs from electric-furnace iron in hardness, assuming comparable analysis.

51. The air furnace permits production of iron of low carbon content and facilitates the introduction of alloys.

52. Ample time is provided in air-furnace melting for preliminary analyses and the introduction of material to correct faulty analysis.

DISCUSSION

MEMBER: Which type of firing does the author use?

D. P. FORBES: We have used both hand firing and pulverized-coal firing, and we have also used oil firing for white iron.

CHAIRMAN JOHN W. BOLTON: Those who have been in the game long enough, remember that years ago practically all the irons above 40,000 lbs. were air-furnace irons. The improvements and changes, methods and manipulations, and greater knowledge that have come to us the last few

* The Lunkenheimer Co., Cincinnati.

years, make it apparent that the air-furnace operators are keeping pace with everyone else.

MEMBER: Mr. Forbes mentioned the use of soda ash in a receiver before the air furnace. Is it possible, when the cupola is connected directly into the furnace, to use the soda ash directly in the furnace itself?

D. P. FORBES: The problem in using soda ash is purely one of getting refractories to stand up in contact with the slag that is formed. We have found after much experimenting, that a good grade of firebrick is the cheapest, and that the smallest area of contact gave the best results. Consequently, we have rather a small receiver and we apply the soda ash at that point. Air-furnace sidewalls, of course, last much longer without any soda ash in the furnace.

JOHN SHAW: * There is one thing in relation to the low-carbon irons which I might bring up. I think it is pretty clearly demonstrated now that superheating irons with carbons much below 3 per cent is of no value; perhaps it is really more of a detriment. Superheating is of more value with the higher-carbon type irons.

R. F. HARRINGTON: † I listened with a great amount of interest to Mr. Forbes' paper on air-furnace practice and concur on a great many of his points, particularly, that the air furnace is a means by which practically any of the higher test types of iron can be produced. It is particularly adaptable to the low-carbon type of iron.

We have demonstrated pretty thoroughly the correctness of Mr. Shaw's statement to the effect that excessive superheating of the lower carbon type of iron is really detrimental, and that much more is to be gained from superheating the irons in the higher carbon ranges. We have also experienced the dendritic structure when the carbons are in the lower range, in those castings of material which we might call special irons.

I found it a little difficult to reconcile some of the data on physical properties that Mr. Forbes gave, with those of our own practice, particularly in connection with the relationship between Brinell hardness and tensile strength. We find that with plain irons running 50,000 lbs. or better in tensile strength, the Brinell hardness is apt to be at least twenty points above the hardness indicated by Mr. Forbes. He showed a transverse load of some 6100 lbs. on the old bar and stated that the tensile would be in the vicinity of 50,000 or 55,000 lbs., showing a Brinell of 196. In our experience with such a tensile result, we would have a Brinell hardness considerably in excess of 200. Of course, with the use of alloys, that hardness could be still further raised.

D. P. FORBES: Briefly in answer to Mr. Harrington, at one time we changed our operation from hand-fired to pulverized-coal firing, and found an increase in Brinell with similar analyses. We did not fire with pulverized coal for any length of time. That might explain, to a certain extent, why Mr. Harrington's results are different from our own.

R. F. HARRINGTON: Operating on pulverized coal in contrast to hand-firing, does not answer the question. As I recall, in the days of the hand-fired furnace we had greater oxidation, and there was a tendency

* Southsea, England.

† Hunt-Spiller Mfg. Corp., Boston.

for a higher Brinell in the case of the hand-fired furnace. However, that may be beside the point.

CHAIRMAN JOHN W. BOLTON: There are two rather significant points that might perhaps be re-emphasized in Mr. Forbes' paper. I believe we should emphasize the point that the factor of time—a greater time allowed in air-furnace operations apparently, according to Mr. Forbes' results—has such an effect that intensity of superheating is not as necessary as in certain other furnace practices.

Another point brought out by the author concerns this dendritic structure, which I believe we may assume to be a result of a decomposition of a primary austenite. It is very interesting to note the results he found in that connection.

MEMBER: What is meant by "superheat?" The word has been used several times in this meeting.

D. P. FORBES: By "superheat," we mean getting the iron hot enough above the solidification point so that it can be handled in the foundry. Probably our iron would come from the furnace at, on the average, above 2700 degs. Fahr., probably at all times less than 2900 degs. Fahr.; that is, as recorded by the optical pyrometer.

CHAIRMAN JOHN W. BOLTON: The average operating temperature of 2800 degs. Fahr. allows one to accomplish the same results in the air furnace that the electric-furnace operator gets with 2900 degs. Fahr.

R. F. HARRINGTON: There is a good treatise on that subject with relation to heavy-ordnance gun manufacture in the middle of the last century, in which it was quite thoroughly demonstrated that the density of the iron and its consequent physical characteristics were very much a function of time and temperature. Those old air-furnace operators were rather limited in those days on the question of temperature because, of course, they did not have the advantage of pulverized-fuel firing. They then used the longer time in the furnace and lower temperature. As the time increased they built up density, and it was by longer time at a lower temperature that they were able to cast 60-ton and 100-ton guns of a density sufficient to stand the firing of heavy powder charges. Thus it is, I believe, very definitely a case of time and temperature; and in the case of the air furnace, the time element is the more important.

The Coreless Induction Furnace as a Laboratory Tool

By WAYNE E. MCKIBBEN,* SCHENECTADY, N. Y.

Abstract

This paper discusses the coreless induction furnace as used for experimental purposes where small quantities of alloy metals are desired. The author first describes the principles of the furnace, then the equipment as used in the General Electric Co. laboratory, also refractories, raw materials, and problems encountered in the production of the various alloys. It has been found that by the use of this type of furnace, steels of widely varying analyses can be produced in sufficient quantities to allow adequate testing, analyses can be controlled with accuracy, and other advantages obtained.

1. A number of papers have been written about coreless induction furnace operation, but in the main they have been concerned with commercial installations involving relatively large furnaces and tonnage production. It is the purpose of this paper to describe the results of five years of operation of 50- and 150-lb. Ajax Electrothermic Corp. furnaces in connection with the Research Laboratory of the General Electric Co.

2. Briefly, the coreless induction furnace is a transformer, the primary consisting of a single-layer cylindrical coil of copper tubing to which electrical energy is supplied in the form of alternating current at a frequency of 500 to 2000 cycles for commercial melting, ten to thirty times the frequency of that utilized for power and lighting purposes. The secondary of the transformer is the charge of metal, or conducting material, which is placed within the coil. Water is forced or drawn through the tubing to carry away the heat generated by the passage of the electric current and conducted through the lining from the charge.

3. The alternating current flowing in the primary causes an alternating magnetic flux to link the coil, and that flux which

* Research Laboratory, General Electric Co.

NOTE: This paper was presented before the round table on steel castings at the 1933 Convention of American Foundrymen's Association.

is within the coil and linking the charge generates an electromotive force which causes eddy currents to flow in the charge, thus heating it. This heating occurs regardless of the nature of the charge, providing that it has an appreciable electrical conductivity, although the rate of heating for any particular frequency and coil and charge diameter is determined by the electrical resistivity and magnetic permeability of the charge and the thermal resistivity of the refractory.

4. Reaction between the electromagnetic field set up in the charge by the coil and the eddy currents in the charge, will cause a fluid charge to circulate or stir, as shown in Fig. 1. The rate of stirring and the height of the meniscus m varies with the frequency, the crucible diameter and the position of the charge vertically within the coil.

5. Since these furnaces operate at higher than normal frequencies, energy is usually supplied from motor generator sets, except for small installations which may be supplied by means of mercury spark gaps at frequencies of 30,000 to 40,000 cycles per second, or by vacuum tubes up to frequencies of 100,000 to 200,000 cycles per second. The power factor of coreless induction furnaces is very low (in the neighborhood of 5 to 15 per cent), so that capacitors are connected in parallel with them, a sufficient number being connected to switches so that the power factor may be adjusted to near unity throughout the melting period.

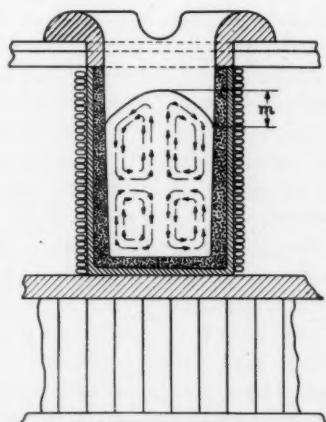


FIG. 1—CROSS-SECTION OF FURNACE, SHOWING STIRRING.

6. A relatively long melting time means a large loss of energy by radiation and conduction from the charge, and a short melting time requires larger and more expensive energy and control equipment. Increased operating frequency increases the difficulty and cost of building the generators and decreases the cost of the capacitor installation. Increasing the size of the charge reduces the minimum allowable operating frequency. At present, the usual procedure is to melt the charge in from one to two hours. The frequency generally used is 960 cycles, although frequencies up to 2000 cycles are in use.

Equipment

7. The equipment which we are using consists of an induction motor-driven, single-phase, 60-k.v.-a., 960-cycle, 440-volt generator with starting, switching and voltage regulating equipment located in the basement of one of the laboratory buildings and connected, by means of a concentric, two-conductor, lead-covered cable, with the furnace room on the third floor, a distance of about 200 feet.

8. In the furnace room the concentric cable terminates at a control panel which carries an oil circuit breaker; control for varying the setting of the voltage regulator; switching equipment for the capacitor bank located immediately to the rear of the panel; an ammeter, power factor meter, and voltmeter connection for the 960-cycle circuit; and a kilowatt-hour meter connected into the motor circuit to indicate the total energy input to the motor-generator set. From this equipment connection is made to a switching panel which allows the use of any of the four positions on the furnace platform (see Figs. 2 and 3).

9. Three of these positions are at trunnions which support furnaces, a 100- to 170-lb. furnace in the center with a 25- to 50-lb. furnace on each side. Each furnace consists of a box of asbestos lumber attached to a brass angle frame with the bottom covered with Sil-O-Cel granules or Sil-O-Cel brick under a refractory cement. The coils are placed upon the cement and held in place by asbestos lumber spacers.

10. Electrical connections are made by means of sliding contacts mounted beneath each furnace within an asbestos lumber enclosure. The contacts are connected to corresponding switches on the switching panel by means of cables which pass through an open-top duct in the furnace platform just to the rear of the fur-



FIG. 2—GENERAL VIEW OF THE MELTING EQUIPMENT.



FIG. 3—CONTROL AND SWITCHING PANELS.

naces, the duct being covered with asbestos boards set flush with the floor. Water connections, consisting of one inlet and two outlets, are made by means of rubber hoses which attach to the water supply and discharge lines, also carried in the duct, and connect to the furnace coils by means of unions on the back of the furnace boxes beneath suitable shields.

11. These arrangements allow rapid interchange of furnaces of which two are available for each of the three positions, thus preventing a stoppage of operations because of furnace failure. City water is used for cooling purposes, since one case of an overheated coil was apparently caused by foreign matter in a supply of poorer quality.

12. A hood connected to a blower, discharging outside the room, is mounted so that it can be swung over the large furnace and one of the small ones to carry away noxious or toxic fumes which may result from the addition of certain alloying elements.

13. The fourth position on the furnace platform consists of electrical and water connections to which any one of several coils may be connected for special uses which will be described later.

Refractories

14. The charge of the coreless induction furnace is deep and of small cross-sectional area, the depth being about 1.25 times the diameter in our furnaces. Furthermore, to obtain satisfactory energy transfer from the coil to the charge, the space between them must be as small as possible in order to obtain good coupling.

15. The first linings which we used consisted of a $\frac{1}{8}$ -in. lining of No. 162 alundum cement inside the coil with a sheet of flexible mica fitted inside this for electrical insulation and for preventing metal, which may seep through the refractory, from reaching the coil. A pre-formed basic crucible with $\frac{1}{2}$ -in. wall, apparently largely magnesia with some less refractory binding material, was placed inside this and the clearance space of about $\frac{1}{2}$ -in. between the mica and the crucible filled with chrome ore tamped lightly into place, a rim and pouring lip being formed on top with refractory cement.

16. Due to the vigorous stirring of the molten charge and the character of the bonding material, the crucible life was quite short. Furthermore, any attempts to dislodge slag encrustations at the metal line were liable to break out a whole section of the crucible wall, allowing the chrome ore to run out and expose the mica and the coil to the heat of the charge and perhaps to the action of the molten metal.

17. After considerable development, a lining was evolved which has proved to be very satisfactory. The alundum cement lining is $\frac{3}{8}$ in. thick, this having been found capable of turning back any metal which might seep through cracks in the crucible wall proper. The mica lining is not used, the increased thickness of the alundum cement having eliminated the need for it.

18. The crucible proper is formed by lightly tamping a basic lining material (82.6 per cent electrically fused magnesium oxide, 12.4 per cent calcined magnesium oxide, 5 per cent "G" sodium silicate) around a tapered graphite plug placed in the coil. No particular attention is paid to the particle size of the refractory. It is ground until it "looks about right," the largest particles being of $3/64$ to $1/4$ in. diameter. (See Table 1.)

19. The lining is then sintered by energizing the furnace

Table 1

GRADING OF A TYPICAL SAMPLE OF REFRACTORY	
Retained by 20 mesh screen.....	10.6 per cent
Retained by 40 mesh screen.....	18.6 per cent
Retained by 60 mesh screen.....	22.6 per cent
Retained by 80 mesh screen.....	5.3 per cent
Retained by 100 mesh screen.....	4.9 per cent
Retained by 120 mesh screen.....	5.0 per cent
Retained by 150 mesh screen.....	3.3 per cent
Retained by 200 mesh screen.....	9.2 per cent
Passed by 200 mesh screen.....	20.7 per cent

Table 2

	Large Furnace, 100 to 170 Lbs.	Small Furnace, 25 to 50 Lbs.
Size of graphite plug, inches.....	15x7 $\frac{1}{2}$ —8 $\frac{1}{2}$ diam.	9 $\frac{1}{2}$ x5 $\frac{1}{2}$ —6 diam.
Weight of graphite plug (finished), lbs.....	45	15
Weight of graphite plug (unfinished), lbs.....	80 (approx.)	25 (approx.)
Kind of graphite.....		Acheson Electrode Graphite
Labor of turning plug, hrs.....	3	2
No. of crucibles fused per plug.....	10	12
Cost of plug per crucible.....	\$1.25	\$0.38
Alundum cement (No. 162), lbs. and cost.....	15	\$1.65
Thermolite cement (for lip), lbs. and cost.....	15	\$0.75
Refractory, lbs. and cost.....	75	\$4.50
Labor for placing, hrs. and cost.....	2	\$1.00
Kw. hrs. for drying cement.....	10	\$0.20
Time required for sintering, min. and cost.....	75—90	50—65
Kw. hrs. required for sintering.....	\$5—100	\$2.00
		45—55
Total cost.....		\$12.10
Maximum no. of heats per lining.....	27*	43
Average no. of heats per lining.....	23—24*	28—30
Cost of lining per heat.....		\$0.50
Cost of lining per lb.*.....	(150-lb. heat)	\$0.0033 (35-lb. heat)
		\$0.0058

*The linings of the large furnace deteriorate considerably from lack of use; otherwise, these figures would be better.

until the plug reaches a temperature of about 2000 degs. Cent. (3632 degs. Fahr.) the proper point being determined by the elapsed time and kilowatt-hour input to the motor-generator. The graphite plug is then removed and cooled in an atmosphere of illuminating gas to prevent excessive burning. After the crucible has cooled sufficiently, a rim and pouring lip are formed as before.

20. The advantages of this type of lining are several. It is formed in place and is quite strong; it presents a smooth, continuous face to the charge from the very beginning of the first heat, with no interstices for the metal to penetrate, and offers no opportunity for erosion before the metallic oxides form a glazed coating; it is highly refractory, and the bonding material is of

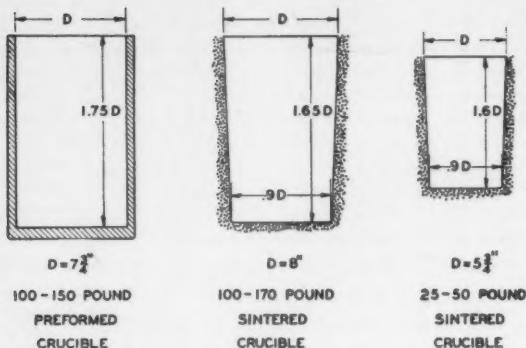


FIG. 4—CRUCIBLE PROPORTIONS.

such a character as to resist the stirring action; and, finally, it is comparatively cheap. (See Table 2.)

21. To return to some of the details concerning the crucibles, the proportions are about as shown in Fig. 4, the burning of the graphite plugs used for fusing altering them somewhat.

22. To facilitate the removal of the graphite plug after the sintering operation, a section 5 inches in diameter and $2\frac{1}{2}$ inches high (for the large plug) is left on the top of the plug through which a $\frac{7}{8}$ -in. diameter hole is drilled transversely. During the sintering operation this top is covered with the refractory used for the lining. When sintering is completed, this top is uncovered, a $13/16$ -in. diameter steel rod 2 feet long is thrust through the transverse hole, pipes 4 feet long slipped over the ends of the rod, and the plug lifted out and placed in a steel tank upon a block of graphite. This arrangement has worked satisfactorily for two years, the only difficulty arising from the occasional melting of the steel rod after the plug is placed in the tank.

23. It will be noted that the crucible life which we obtain is rather poor as compared with that usually reported. However, in view of the intermittent and variable services required, these figures are not so bad.

24. Never more than a few heats (usually only one) are poured at a time, the crucibles being heated and cooled quite frequently with the consequent expansion and contraction which cause cracking. Furthermore, the furnaces may be idle for several days at a time with the crucibles absorbing moisture, which

is evaporated quite rapidly when a heat is made, with further stress upon the lining.

25. About 15 per cent of the heats which we make are of high silicon content, *i.e.*, 3 to 7 per cent, which causes considerable erosion of the crucible. Successive heats quite frequently vary markedly in composition, so that it is necessary to make frequent wash heats (about 25 per cent) of high silicon content with occasionally the addition of considerable fluorspar to scour out the crucible, thus still further reducing the crucible life.

26. The type of basic lining described is the one we generally use. For the occasional heats requiring an acid lining we use a pre-formed crucible similar to the basic one first described. The life which we obtain from these is usually very short, only 4 to 6 heats. We have also used zirconium silicate as furnished by the furnace manufacturers in place of the magnesium oxide. The life seems fairly good, but insufficient data are available to warrant a prediction of the life which they will give.

Raw Materials

27. Since most of the melting that we do is developmental work, it has seemed desirable to use commercially pure raw materials to eliminate as many of the variables as possible. For the iron in our charges, we have used 1½-in. to 2-in. Armco squares ¼ to ⅜ in. thick. These melted quite readily and were convenient to use, but the melts required considerable deoxidation.

28. Later, a deep-drawing scrap of the following analysis became available: carbon 0.05, manganese 0.2, phosphorus under 0.02, sulphur 0.028, silicon a trace. This has been in use ever since and has proved very satisfactory. It melts down quietly and requires very little deoxidation.

29. An important factor in its good performance is the fact that it can be obtained in the form of 6-in. diameter disks about 3/32 in. thick. These fit easily into the large furnace, giving a very good space factor and good coupling between the charge and the coil, thus reducing the melting time. The latter, coupled with the reduction of the surface-volume ratio resulting from careful stacking of the disks, is of importance in reducing the deoxidation required.

30. For nickel alloys we have achieved best results using 1½-in. to 2-in. squares of electrolytic nickel. Shot nickel seems to melt down with the evolution of considerable gas.

31. Cobalt is used in the form of rondles, $\frac{1}{2}$ in. diameter, $\frac{1}{2}$ in. long. This material as received from the importers contains considerable oxide, as shown by the condition of the interior of the pieces. Although we have occasionally cleaned this up by hydrogen firing, we do not resort to this procedure unless a shipment is very bad or the requirements of a particular melt quite severe.

32. Chromium is used in two forms, 60 to 70 per cent ferro-chrome with carbon less than 0.3 per cent, and chromium metal containing about 2.5 per cent iron and 0.25 per cent carbon.

33. The manufacture of X-ray and vacuum tube parts in the laboratory results in considerable amounts of very pure molybdenum and tungsten scrap, all of which has been refined in the laboratory prior to working. Our supply of these metals is obtained from such scrap. Quite frequently they are in the form of chunks of such size that they can be dissolved only by prolonging the melting period considerably.

34. Low-carbon ferrosilicon and ferromanganese of approximately 90 per cent alloy content are used. For other alloying elements, such as titanium, vanadium, zirconium, high-grade ferroalloys are used.

Melting Procedure

35. In connection with all melting in the coreless induction furnace, the factor of the vigorous stirring previously mentioned must be kept in mind. In the case of our large furnace, it is so vigorous as to cause the surface of the melt at the center of the crucible to be 2 to 3 inches above the surface at the walls. This means that there is considerable opportunity for oxidation during and after the melting period unless the melt temperature is kept low.

36. This tendency is of considerable consequence when part of the charge wedges against the walls on the way down and the molten portion at the bottom becomes overheated. Under such circumstances the melt usually becomes very wild and causes considerable difficulty. We have found it advantageous to allow such melts to solidify when they are completely melted down, causing a considerable amount of gas to be evolved. Upon remelting, the melts are usually found to be more tractable.

37. In addition to melting as rapidly as possible and keeping the temperature low, we ordinarily add 0.2 to 0.4 per cent manganese, one-half when the charge is about half melted, the other

when it is nearly all down. This, in connection with the 0.2 per cent manganese in the iron scrap, keeps the melt in good shape. The final manganese is 0.2 and 0.4 per cent.

38. Shortly after the last manganese addition, about one-half of a 0.2 to 0.3 per cent silicon addition is made, the melt warmed, power cut off, and any slag skimmed from the surface with an iron rod. Power is then put on, final additions made, followed by the other half of the silicon, proper temperature attained, and the melt poured as quickly as possible. If there is little slag after the first silicon addition, the power may be put on immediately without skimming, final additions made, and the melt poured without the last half of the silicon addition, providing the time taken for these operations is not too long.

39. This general procedure is varied somewhat with certain types of steel to give the best results. In the case of high-silicon steels (3 to 15 per cent), we melt down as described previously, except that we keep out a small part of the iron scrap, make the complete silicon addition just after the last manganese addition, and follow this with the remaining scrap to cool the melt to the skimming temperature. If the skimming operation does not take too long, we pour at once without reheating. If additions other than the silicon are to be made, we make them with power on after the skimming operation, adjusting the power input so as to have the correct temperature for pouring by the time the last additions have been sufficiently distributed.

40. For melts containing more than 5 to 10 per cent nickel, we have found it desirable to melt very carefully and to reduce the amount of deoxidizers used as much as possible. The nickel is charged beneath the iron scrap and one-half of the manganese addition made when the charge is half down, the remainder at end. Final manganese is about 0.6 per cent.

41. Considerable difficulty was encountered with the addition of cobalt because of the evolution of gas. This is now overcome either by mixing the cobalt with the iron of the charge, or by placing it in the crucible on top of the iron in such a position that it will have a chance to heat thoroughly before it reaches the molten metal.

42. Due to the size of the pieces of tungsten scrap, we have occasionally found it necessary to place them on the bottom of the crucible beneath the remainder of the charge to allow them sufficient time to dissolve. Using this procedure, we have had no

difficulty in making alloys containing as high as 30 per cent tungsten.

43. The making of a plain carbon steel gave us considerable difficulty for quite some time. With no melting stock containing appreciable carbon available, and with the probability that the recovery would be quite small if it were used, we chose to add carbon to the completely molten charge by means of wash metal.

44. Regardless of the melting procedure used, excessive boiling always occurred whenever even a small amount of wash metal was added. This frequently caused a crucible half full of metal to boil over, and very frequently caused "bridging" to occur quite a distance above the melt, making it necessary to hammer the metal until it was broken away and returned to the melt, with consequent damage to the lining.

45. Since only small pieces of wash metal could be added at a time, because of this boiling, considerable carbon was lost through oxidation during the time required for the wash metal addition. Altogether, the results were quite erratic and unsatisfactory.

46. After trying the addition of carbon by means of graphite placed on top of the melt, which gave 20 to 40 per cent recovery, we finally hit upon this scheme which has been much more successful. We melt the charge as usual with the manganese additions at the half-way and end points, holding out 5 to 10 per cent of the iron, add about 0.2 per cent silicon, plunge the carbon addition (graphite rods) to the bottom of the melt, folded in a flat cold rolled bar, and hold it there while the remaining iron scrap is added to solidify the melt on top.

47. In about 5 minutes power is put on at about half normal voltage, and the voltage increased in 5 to 10 per cent steps at 5 to 10 minute intervals until the charge is molten. This causes the graphite to dissolve before the top of the melt is broken through.

48. As soon as the charge is remelted, about 0.1 per cent silicon is added and the melt poured. If there are further additions, they are made after the second silicon addition.

49. This is perhaps a trick method of melting, but it has worked exceptionally well for us. The final carbon is approximately 10 per cent under the addition, and the results are very uniform. Furthermore, we have no trouble with excessive boiling, spitting and bridging.

50. With the stirring and consequent oxidation which occurs with this furnace, it is clear that some adjustment of additions

Table 3

Nickel 20% and over	—add 0.25-0.5% to desired per cent.
Cobalt 20% and over	—add 2% to desired per cent.
Chromium	—no increase.
Carbon	—increase by 1/10—recovery 90%.
Vanadium up to 1%	—increase by 1/2 —recovery 67%.
Vanadium 1 to 3%	—increase by 1/4 —recovery 80%.
Silicon up to 7%	—increase by 1/10—recovery 90%.
Silicon over 7%	—increase by 1/20—recovery 95%.
Manganese	—no increase for small percentages.

Table 4

	Charge Wt., lbs.	Ingot Wt., lbs.	Per cent Recovery.
Silicon steel.....	140	136	97.2
Highly alloyed tool steel.....	170	161	94.7
Highly alloyed tool steel.....	165	162	98.2
Highly alloyed tool steel.....	average of 6 ingots		97.2
Nickel-chromium alloy.....	140	135 $\frac{1}{2}$	96.8

must be made to obtain the correct analysis in the ingot. The amount of increase is dependent upon several factors, first, whether melting in the large or small furnace, the iron scrap having good coupling in the large one and stacking up so as to reduce oxidation; second, the other elements which are added; and, finally, the melting procedure.

51. The figures of Table 3 should, therefore, be considered as the increases which are necessary under our conditions. Some idea of the total melting loss can be obtained from the figures of Table 4.

52. The time required for melting varies with the type of material, of course. Using a charge made up of one-half disks of deep-drawing scrap previously mentioned, the other half of alloying elements of various shapes and sizes, and starting with a hot furnace, we have been able to melt and pour a 170-lb. charge of tool steel in 1 hour 3 minutes with an energy consumption at the motor terminals of 835 kw.-hrs. per ton of charge and 883 kw.-hrs. per ton of ingots. Ordinarily, such melting requires from 1 hour 15 minutes to 1 hour 30 minutes, and the energy consumption is 975 kw.-hrs. per ton of charge and 1000 kw.-hrs. per ton of ingot.

53. In connection with these figures, it should be remembered

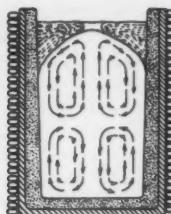


FIG. 5—EFFECT OF STIRRING ON SLAG.

that they are for a 150-lb. furnace and a very satisfactory type of scrap.

54. Temperature measurements by means of optical pyrometers have not proved satisfactory, so that we usually resort to the "rod test," which is crude but quite effective. We thrust a 3/16-in. iron welding rod into the melt and note the speed at which it melts. When the temperature appears to be correct, we count the seconds required to melt it to a certain shape, usually with the power on, although it may be done as easily with the power off if the rod is stirred slightly and allowances made for the difference between rapid stirring and quiescence. This method has served quite satisfactorily, and we have been able to duplicate results quite readily.

55. In order to eliminate the considerable deoxidation which we found necessary at first, and also to make possible some refining action, we tried using slags over our melts. However, the slag tended to be carried away from the center by the circulation of the metal, so that a continuous slag cover could be maintained only with difficulty. (See Fig. 5.)

56. Increasing the amount of slag was only partially successful, because so much material had to be added to keep the melt covered that its removal before pouring was quite difficult. Furthermore, the upper surface of the slag tended to solidify when such a large amount was used, and so increased the difficulties of handling.

57. Reducing the stirring by reducing the power input or by raising the upper surface of the melt with respect to the coil, either by adding an excess of metal or by raising the crucible, was not successful. When the stirring in the upper part of the melt had been reduced sufficiently to avoid carrying away the slag, the input to the upper part was insufficient to keep it molten. A rim

of metal froze around the crucible so that the slag had little opportunity to react with the metal. Rapid solidification occurred when the power was entirely cut off, due to heat loss through the thin refractory to the water cooled coil, so that such procedure was impossible.

58. We have concluded that within the combination of our frequency and crucible diameter (*i. e.*, 960 cycles and 8 inches, respectively), we cannot reduce the power input at the top by an amount sufficient to make the stirring satisfactory without danger of solidification occurring there. At present a further investigation regarding this is being made.

59. One of the things which we desired to do was to hold a melt under a lime slag in a deoxidized condition to allow the elimination of sulphur. In attempting this we added a large amount of slagging material and stirred it well with a small iron rod to prevent uncovering the center of the melt and the solidification mentioned previously. By overheating the melt considerably, sufficient time for the removal of the large amount of slag was allowed while the melt cooled to the pouring temperature.

60. Only with the greatest care were we able to obtain and maintain a white slag for any length of time (10 minutes or so), and then only by means of a somewhat dubious expedient, as outlined below.

61. The melt was slagged with a 50-50 mixture by weight of lime and alumina, to which powdered aluminum was added from time to time (quite a considerable amount altogether), each aluminum addition being covered with some of the lime-alumina mixture and a hydrogen atmosphere being maintained over the whole. Whether the difficulty was due to oxides which the melt absorbed from the crucible, or to oxide already in the melt from the raw material and the oxidation which occurred during the melting down, or to the relatively small amount of slag, has not been determined.

62. The procedure was hardly satisfactory because it resulted in ingots with high aluminum content and because it took a great deal of time. In addition, it brought little improvement in the steel. Because of the difficulties attending slagging operations in our furnaces, they were discarded in favor of rapid melting of selected scrap.

63. In addition to our regular melting operations, we have used our equipment for a variety of special work. We produced

very pure iron by melting a charge of Armeo iron under hydrogen and then bubbling hydrogen through the molten metal for about 20 minutes. Pouring was carried out under hydrogen also. Although the analysis of this ingot (carbon 0.01, manganese 0.009, phosphorus and sulphur very slight trace, silicon a trace) was not especially noteworthy, the magnetic properties were very good and indicative of a high degree of purity.

64. By melting copper scrap under an atmosphere of hydrogen, superheating it for a time, and pouring it into a mold at the right temperature, we have been able to pour good quality ingots of oxygen-free copper, having conductivities of as high as 102.6 per cent. (A.S.T.M. standard is 98.5 per cent.)

65. Using graphite containers packed in powdered graphite, we have been able to reach and maintain high temperatures throughout a relatively large volume. A temperature of 2000 degs. Cent. can be reached in about 30 minutes, and 2200 to 2500 degs. Cent. in a somewhat longer time. These temperatures may be held for several hours without difficulty.

66. Occasionally we use induction heating for relieving strains in partially completed spinnings. This requires only 2 to 7 minutes and enables subsequent spinning operations to be carried out without delay. For certain types of laboratory annealing of magnetic materials, this type of heating also offers advantages.

67. Induction heating has proved far superior to other means in the assembly of large vacuum tube parts. These are made of copper and silver soldered together at a temperature of 775 degs. Cent. The use of torches was unsatisfactory because of physical discomfort, excessive oxidation, and the time required. By placing the parts in a spring-tightened jig made of asbestos board and small brass rods, and placing the whole in a glass bell jar which could be evacuated, we performed the operation without the difficulties previously experienced. Due to the fragility of the glass seals, 15 to 20 minutes was required for the operation, which otherwise could have been performed in about 5 minutes.

Conclusion

68. In conclusion, a word as to the general qualifications of such a furnace for laboratory work. We have found that by its use, steels of widely varying analyses can be produced in sufficient quantities to allow adequate testing.

69. Ingots can be produced upon short notice, 1 to 2 hours

if necessary. Analysis can be controlled with accuracy. The bath temperature can be raised or lowered quickly to suit conditions, since the heat is generated within the charge itself and the amount of refractory is small. The atmosphere over the melt may be controlled with little difficulty. In short, most of the variables are easily controlled.

70. Finally, since two men can pour 400 to 600 lbs. of steel per day, small production requirements of special alloys or special castings may be adequately met.

A Practical Foundry Test on the Effect of Phosphorus, Aluminum and Silicon on Leaded Bronze

By HAROLD J. ROAST,* F.C.S., F.C.I.C., M.E.I.C.,
MONTREAL, P.Q., CANADA

Abstract

To determine the effects of small quantities of phosphorus, aluminum and silicon on the fracture and outward appearance of leaded bronze castings, experiments were run on scrap metal and virgin metal mixtures. The scrap metal was poured into journal bearings, and the virgin metal into large locomotive drivers. Various pouring temperatures and percentages of additions were tried out. Conditions of the resulting castings were compared. Test data convinced the investigators that aluminum and silicon must be kept out of the mixtures used to give best results.

1. The object of the investigation described in this paper was to find out the effect of small quantities of phosphorus, aluminum and silicon on the fracture and outward appearance of leaded bronze castings. The reason for the interest in this matter was the fact that a brass foundry making castings of leaded bronze from customers' scrap found that some of these were unsatisfactory as to their general appearance and fracture. The castings contained from 7 to 8 per cent tin, 21 to 15 per cent lead, and about 2 per cent zinc, the balance being copper.

2. In view of the frequent use of phosphorus, aluminum and silicon in special bronzes, it seemed possible that these elements might be present in the scrap used. It was decided, therefore, to make an investigation into the effect of these elements on the castings produced.

* Lecturer, Department of Metallurgy, McGill University.

NOTE: This paper was presented and discussed before one of the nonferrous sessions at the 1933 Convention of American Foundrymen's Association.

3. It is known that the general appearance and fracture of a bronze casting is affected not only by its composition but also by the size and shape of the castings, the temperature at which it is cast, and by other details of foundry practice. Consequently, it was thought best to conduct the tests under regular foundry conditions rather than those which might seem more scientifically satisfactory in the testing laboratory.

4. Whether this increases or decreases the value of the investigation will depend largely on the point of view of the individual. The writer is intimately acquainted with both lines of investigation and believes that the ideal way of looking for the solution of any foundry problem is the happy combination of high-class scientific research and practical foundry experience.

**EXPERIMENT NO. 1—JOURNAL BEARING CASTINGS
GROUP 1 (CASTINGS 1-8, INCLUS.)**

Base Metal Used for Tests.

5. The tests on journal bearings were started by making up a heat of bronze as follows:

	Lbs.	Per Cent.
Clean cabbaged copper wire.....	245	70
Straits tin	21	6
Tadanac lead	73½	21
Tadanac zinc	10½	3
 Total.....	350	100

Melting Practice for Base Metal.

6. A new plumbago crucible was put in a crucible pit furnace fired with city gas. When the crucible was at a good red heat, a small shovel of crushed charcoal was thrown in and the copper added, then more charcoal placed on top. When the copper had melted and was superheated to about 2200 degs. Fahr., the lead was added, followed by the zinc and, lastly, the tin. The crucible was kept covered with charcoal and the temperature brought to 2300 degs. Fahr., when the crucible was removed from the furnace, the metal stirred well, allowed to cool to 2200 degs. Fahr., stirred again, and the metal poured into molds of 5 x 9 A.R.A. journal bearings (Castings Nos. 1 and 2).

7. These and all later castings made were numbered from 1 to 34, each number denoting some difference either of temperature of pouring or composition, or both.

8. The remainder of the metal was divided by pouring half into another hot pot. The two pots then were put back in the furnace and 1 lb. of phosphor copper (phosphorus content 15 per cent) added to one pot, $\frac{3}{4}$ lb. aluminum to the other. The two pots were brought up to 2300 degs. Fahr.

9. Half the phosphor alloy was poured at 2200 degs. Fahr. (No. 3) and half at 1900 degs. Fahr. (No. 4). Half the aluminum alloy was poured at 2200 degs. Fahr. (No. 5) and the other half at 1900 degs. Fahr. (No. 6). Another original heat of 350 lbs. of journal metal made up as in the first instance was cast at 2200 degs. Fahr. and gave normal castings (no number was given this casting).

10. To the remainder, after casting the test journal bearings, $2\frac{1}{2}$ lbs. of silicon copper (10 per cent silicon) was added, and the metal cast at 2200 degs. Fahr. (No. 7) and at 1900 degs. Fahr. (No. 8).

Table 1
ANALYSIS OF FLOCCULENT POWDER OBTAINED IN POURING JOURNAL BEARING
METAL WITH SILICON ADDED

	Per Cent.
Silica	18.65
Lead	8.69
Iron and alumina	1.42
Zinc	60.30
Copper (by difference)	10.94
	100.00

11. When the metal came to be poured after the addition of silicon, no difference in the surface of the molten metal was discernible; but on pouring into the mold both at 2200 and at 1900 degs. Fahr., a very peculiar light greenish-grayish-whitish powder swirled up in the gate and finally settled around the edge of the gate. This is the most characteristic reaction obtained so far, as neither phosphorus nor aluminum produced it.

12. This flocculent powder was found to contain fine globules of copper-colored metal mixed with a whitish powder. The weight of the sample obtained was 0.3964 gram, from which was obtained 0.2384 gram of metal as referred to above. This left 0.1580 gram as the true weight of the fluffy powder, which was analyzed with the results shown in Table 1. From this analysis

the extreme effect of silicon can be appreciated, oxide and silicate of zinc being produced.

SUMMARY OF EXPERIMENT NO. 1
GROUP 1 (CASTINGS 1-8, INCLUS.)

13. Returning to the heat of metal, we have castings of four compositions, Nos. 1 and 2 the original heat, 3 and 4 original metal plus phosphorus, 5 and 6 original metal plus aluminum, and 7 and 8 original metal plus silicon, all cast at 2200 degs. Fahr. and 1900 degs Fahr., respectively.

Surface Appearance.

14. Castings Nos. 1 and 2 had the regular appearance of sand-cast journal bearings. There were a few small holes on the back due to a slight wash of sand in the mold, but not more than is customary for a good casting. The color of the surface was brownish red, with no sign of a leady color.

15. Casting No. 3—Rich reddish-golden surface, somewhat inclined to gray where sand adhered, but otherwise the same as Nos. 1 and 2.

16. Casting No. 4—Extremely glorified reddish-golden surface, otherwise the same as No. 3.

17. Casting No. 5—The whole surface of the bearing was silvery white, and at some places the top was wrinkled.

18. Casting No. 6—Same as No. 5.

19. Casting No. 7—The surface of the casting was completely covered with a white powder that adhered sufficiently to come off in flakes. On brushing off the surface, it was found to be deeply corroded, having a worm-eaten structure. After standing two days the cleaned surface turned distinctly green.

20. Casting No. 8—Surface similar to No. 7.

Appearance of Fracture.

21. Casting No. 1—Broken rather warm. Had a golden color more inclined to yellow, possibly due to having been fractured while warm. Showed usual chill line of gray and also a uniform fine crystalline fracture in general without any dendritic crystals.

22. Casting No. 2—Broken when cold. This gave the usual chill line, with grayish color for general fracture, devoid of dendrites and free from uneven color.

23. Casting No. 3—The gray chill effect was greater than

in Nos. 1 and 2. In the center, both as to length and cross-section, a golden streak developed indicating incipient intercrystalline shrinkage. Some of these bearings had only golden spots without the suggestion of separation.

24. Casting No. 4—The general fracture was of a gray color, except in the center where for about two inches the extreme golden-colored area existed. This colored area was homogeneous, not intermixed with gray crystals, and there was no sign of a dendritic structure running at right angles to the length of the bearing.

25. Casting No. 5—The fracture showed a much greater depth of chill than the original metal (Nos. 1 and 2), and the usual leady color. Down the center was a lemon-colored area entirely different from the fracture with phosphorus in Nos. 3 and 4. The crystals seemed only loosely held together, although not showing the actual presence of dendrites to the naked eye.

26. Casting No. 6—The lemon-yellow colored area started more closely from the inner surface and ran almost the whole length as well as across the whole width of the journal, and showed practically complete shrinkage detachment between the yellow portion and the grayer area due to chill. In other words, there was exaggerated intercrystalline shrinkage, but still none of the dendritic structure running at right angles to the cross-section was evident.

27. Casting No. 7—Fracture showed a fine mixture of golden-colored crystals interspersed with gray and some dirty brown. The center of the journal had a concentrated dark area about two inches long.

28. Casting No. 8—Similar to No. 7.

EXPERIMENT No. 1, GROUP 2 (CASTINGS 9, 10, 11 AND 12)

29. Continuing the experiments, the phosphorus metal was divided into two parts and equal quantities of aluminum and silicon metal added, respectively. The phosphorus-aluminum metal was cast into journal molds at 2150 degs. Fahr. (No. 9) and at 1900 degs. Fahr. (No. 10). The phosphorus-silicon metal was cast into similar molds at 2200 degs. Fahr. (No. 11) and at 1900 degs. Fahr. (No. 12).

Surface Appearance.

30. Casting No. 9—Rather bluish-reddish color, tending to silvery. No wrinkles.

31. Casting No. 10—Similar to No. 9.
32. Casting No. 11—Evidence of white deposit or incrustation but not quite so pronounced as for silicon alone (some silicon probably burnt out). No worm pits.
33. Casting No. 12—Similar to No. 11.

EXPERIMENT No. 1, GROUP 3 (CASTINGS 13 AND 14)

34. Continuing the experiments, equal parts of the phosphorus-aluminum and phosphorus-silicon metals were melted together and cast at 2200 degs. Fahr. (No. 13) and at 1900 degs. Fahr. (No. 14)

Surface Appearance.

35. Both castings Nos. 13 and 14 were silvery in color, with no worm holes.

Fracture Appearance.

36. Casting No. 13 had a few golden areas, but otherwise the fracture was normal. There was no discolored area usually attributed to "steam flash."

DISCUSSION OF RESULTS OF EXPERIMENT No. 1

37. Taking the whole of these experiments into consideration, in which phosphorus at 0.08 per cent, aluminum at 0.49 per cent and silicon at 0.08 per cent were added to the journal metal, the indications are as follows:

- (a) That phosphorus alone is the least inimical to good surface appearance and good fracture of the casting.
- (b) That aluminum affects the outside of the journal casting, producing a silvery-white color, and tends to spoil the fracture especially at the higher casting temperature.
- (c) That phosphorus and aluminum are worse than aluminum alone.
- (d) That silicon, or phosphorus and silicon, both destroy the surface of the casting as well as color the surface dead white, but do not affect the fracture to the same extent.
- (e) That continued remelting will largely eliminate phosphorus and silicon, and to some extent aluminum also.

EXPERIMENT NO. 2—CHECK TESTS ON JOURNAL BEARINGS
WITH SMALLER PERCENTAGES OF P, AL AND SI
(CASTINGS 15-20, INCLUS.)

38. Following upon the foregoing work, it was thought well to repeat the tests using only 0.05 per cent phosphorus, aluminum and silicon, and so not only check the first findings but also see whether smaller amounts would still give definite reactions. To this end, 350 lbs. of synthetic journal metal was made up and journal bearings cast at 2300 degs. Fahr. (No. 15), the balance being cast at 1900 degs. Fahr. (No. 16)

39. Casting No. 15—The surface appearance in general was satisfactory.

40. The fracture, however, was not gray but light golden and showed a line of actual separation somewhat near the upper surface of the fracture running longitudinally for two-thirds of the length of the bearing. This evidently is due to the fact that the chilled portion of the outside of the casting is stronger; in the tearing and shock effect of the hammer blow, the weaker large dendritic formation is separated from it. These dendrites do not appear to be at right angles to the length of the fracture (as is said to be characteristic of the silicon fracture), but in the inside of the fracture large dendrites appear which would be very difficult to distinguish from those due to silicon.

41. The moral is that good metal, free from phosphorus, aluminum and silicon, if poured at a sufficiently high temperature, will give a fracture that is in the main golden-colored and has no steam flash but will separate under shock, and will give evidence of dendritic structure very similar to that produced by silicon.

42. Casting No. 16—The surface appearance of these castings was quite satisfactory. The fracture was gray throughout, free from any sign of separation, and contained no golden areas and no evidence of steam flash, brown or otherwise. This emphasizes the importance of temperature in making castings.

43. One-third of the original alloy was then taken and an amount of phosphor-copper equivalent to 0.05 per cent phosphorus was added and journals cast at 2200 degs. Fahr. (No. 17) and at 1900 degs. Fahr. (No. 18) The fractures in both cases were of a gray color with little golden crystals interspersed, but there were no large central areas and no steam flash.

44. To another one-third of the original metal an amount

of silicon-copper equivalent to 0.05 per cent silicon was added. The first castings were made at 2200 degs. Fahr. (No. 19) and the others at 1900 degs. Fahr. (No. 20) As in the case of 0.08 per cent silicon, a light fluffy material showed up in the gate; on shaking out, the characteristic white zinc oxide covered the whole bearing, and a very pronounced worm-eaten structure was on the surface. The fractures in both cases were passable and were similar in appearance to those obtained when the larger amount of silicon was added, although to a lesser degree.

45. From this second experiment it is evident that phosphorus up to 0.05 per cent may not give any trouble, although it is the writer's opinion that it is better kept around 0.02 per cent. It is evident that 0.05 per cent aluminum is absolutely detrimental to the appearance of the casting and dangerous as to the fracture, and that the same may be said of 0.05 per cent silicon. These conclusions are based on journal metal of the composition stated. The same additions to alloys of different composition might well have different results.

EXPERIMENT No. 3—LARGE LOCOMOTIVE DRIVER

46. With the results of the previous experiments in mind, a similar line of investigation was undertaken on the effects of temperature and of additions of 0.05 per cent phosphorus, aluminum and silicon, using metal of the following composition:

	Lbs.	Per Cent.
Tadanac copper.....	262½	75
Straits tin.....	28	8
Tadanac lead.....	52½	15
Tadanac zinc.....	7	2
<hr/>	<hr/>	<hr/>
Total.....	350	100

47. In place of the A.R.A. journal pattern, a pattern for a large locomotive driver, cast weight 300 lbs., was used.

Melting Practice for Base Metal.

48. The same melting procedure was followed as in the other heats, one driver being cast at 2100 degs. Fahr. (No. 23), the other at 1900 degs. Fahr. (No. 24) These were left in the sand one hour, then removed, the sand brushed off, and cooled for another hour on the floor. They then were put outside for one hour, where the temperature was 32 degs. Fahr., and then broken.

49. These castings were uncomfortably warm to the hand,

but there was no evidence of oxidation of the fracture. No. 23 had a well-defined chill to a depth of $\frac{1}{8}$ to $\frac{1}{4}$ inch, and a uniform cross-section for the remainder consisting of light golden crystals intermixed with leady-colored crystals. There was absolutely no sign of intercrystalline shrinkage or longitudinal separation of dendrites. Strain fins showed in three or four places, but there was no actual swelling of the casting. Casting No. 24 had a similar appearance to No. 23, except that the proportion of golden crystals was less. The casting showed no sign of shrinkage.

Addition of 0.05 Per Cent Phosphorus.

50. These castings and gates were put back in the pot and heated to 2200 degs. Fahr. and 0.05 per cent phosphorus added as 15 per cent phosphor-copper. Similar castings were made at 2100 degs. Fahr. (No. 25) and at 1900 degs. Fahr. (No. 26)

51. Little difference, if any, was noticeable in the color of the fractures as compared with Nos. 23 and 24 made from virgin metal, except that No. 25 ate a little more into the sand, although only slightly so. There was no tendency for the castings to swell out at the base, nor any sign of shrinkage on the surface of the casting. In other words, the 0.05 per cent phosphorus or less (some may have been used up in deoxidizing) is not a critical matter in the making of drivers.

Check on Casting Temperature Effects.

52. Two more pots of original metal were made up of the same composition as before. One was brought to 2400 degs. Fahr. This pot was in the furnace for one hour after it had reached a temperature of approximately 2200 degs. Fahr. It had been covered with charcoal most of the time, but at the last the charcoal had burned off.

53. The same engine driver pattern was used and the first mold cast at 2350 degs. Fahr. (No. 27) On removal from the sand, this casting showed some strain fins but no swelling and no shrinkage. The fracture was satisfactory, the usual chill of $\frac{1}{4}$ inch or so was noticeable, and the general fracture was that of light yellow crystals with a few leady-colored crystals intermixed. Very fine holes were evident, generally distributed. There were no dendrites and no sign of incipient shrinkage.

54. The second mold was cast at 1900 degs. Fahr. (No. 28) This casting was free from strain fins and showed no sign of general shrinkage. The general fracture looked very similar to

that of No. 27, except that it was more ready in color; there were very few gas holes, and those very minute. Both Nos. 27 and 28 would be considered entirely satisfactory castings.

55. The second pot was brought up to 2350 degs. Fahr. and cast at 2300 degs. Fahr. for the first mold (No. 29), and at 1900 degs. Fahr. for the second mold (No. 30). Both castings were similar to Nos. 27 and 28 as to appearance and fracture, except that No. 29 showed a slight indication of incipient orange spots in the fracture, but no separation or dendrites.

Addition of 0.05 Per Cent Aluminum.

56. These four castings were then put back into their respective pots, making two pots in all. The metal was brought up to 2300 degs. Fahr.; 0.05 per cent aluminum was added to one and the mold cast at 2300 degs. Fahr. (No. 31), and the other mold cast at 1900 degs. Fahr. (No. 32)

57. Casting No. 31 showed no sign of silvery white on the surface, and no sign of shrinkage in the casting itself. Strain fins were very evident, due probably to the high temperature of casting and possibly to slightly lighter ramming of the sand. The fracture, however, was very unsatisfactory, showing in color from dull red to dull brown. Large areas of dendrites loosely held together were evident to the naked eye, being $\frac{1}{8}$ inch long. There also was a pale golden-colored area having a smooth surface.

58. Casting No. 32 showed no sign of silvery-white appearance on the surface, and no strain fins. The fracture showed several shrinkage holes having a smooth golden-colored surface. The longitudinal fracture contained a bad area in the center where the color was golden to reddish brown, with distinct shrinkage separation to a length of about two inches.

Addition of 0.05 Per Cent Silicon.

59. The metal remaining in the pot was brought to 2300 degs. Fahr. and 0.05 per cent silicon added in the form of 10 per cent silicon-copper. The first mold was cast at 2200 degs. Fahr. (No. 33) and the second at 1900 degs. Fahr. (No. 34)

60. Casting No. 33 showed a dead-white surface on the length of the gate all around, and the casting had a good many strain fins but no worm-like corrosion. The fracture showed considerable intercrystalline shrinkage and a tendency to separation for a length of about two inches. There was a dark orange spot along the center line of the casting. The usual chilled border was in

evidence with dendritic crystals running at right angles to the fracture from the chilled area toward the top of the casting for about one-half inch.

61. Only one fracture showed the foregoing, the remaining pieces giving no evidence of such columnar dendritic structure. The fracture in general was of light yellowish color, with fine gas holes and one or two large shrinkage holes.

62. Casting No. 34 showed the characteristic white appearance on the surface, with very heavy worm-like corrosion especially in two or three large areas. There were no strain fins and no general shrinkage in the casting itself. The fracture of two pieces was quite normal, but one piece showed a fracture having a segregated area of light yellow crystals in the center. There was no sign of actual separation.

GENERAL CONCLUSIONS

63. To sum up, the addition of 0.05 per cent phosphorus to metal used for engine drivers gave no evidence of trouble as to surface appearance or fracture when cast at high or low temperatures, although the lower temperature did have a closer fracture, as would be expected. At the high temperature, however, the metal did not eat into the sand, although there were a few strain fins.

64. The addition of 0.05 per cent silicon to the same metal gave evidence of serious defects at both high and low temperatures.

65. The addition of 0.05 per cent aluminum gave no evidence of its presence on the surface of the casting but showed clearly in the fracture, giving serious dendritic structure at the high temperature and still more internal shrinkage separation at the low temperature.

66. It is interesting to note the difference in the action of the additions to the two alloys used.

Phosphorus. The addition of not more than 0.05 per cent phosphorus is not injurious to either alloy, but there is greater need to keep the casting temperature as low as practicable.

Aluminum. The addition of 0.05 per cent aluminum to the journal bearing alloy produced a silvery-white surface, but the same amount added to the engine driver metal gave no evidence of such color on the surface. In both alloys, however, the aluminum spoiled the fracture.

Silicon. The addition of 0.05 per cent silicon was evident in both alloys as producing a white surface, but in the case of the engine driver metal the low-temperature castings showed no trouble in the fracture, and the high temperature no corrosion pits.

67. The result of the tests of the 34 castings examined has convinced the foundry that aluminum and silicon must be kept out of these particular alloys, at least when satisfactory castings of the type indicated are to be made. With the advent of silicon bronzes and the increased use of aluminum alloys in many brass foundries, the difficulty of preventing an admixture of the harmful elements is considerable.

68. The author again wishes to emphasize the point that this investigation was intentionally confined to foundry operations. If it is of any assistance to any foundry superintendents or foremen as showing a general plan of attack on some foundry problems, the work done will have been worth recording.

DISCUSSION

H. J. ROAST: The real purpose of this paper lies in the first four words: "A Practical Foundry Test." It is addressed primarily to those superintendents and foremen of nonferrous foundries who from time to time find themselves faced with the problem of rejected castings after they have done all they know how to obtain satisfactory castings.

MEMBER: Would the author explain his reasons for arriving at the conclusion, in the last sentence of paragraph 12, that this particular powder was oxide and silicate of zinc?

H. J. ROAST: In that connection, I think I stand open to criticism, in that I did not pursue the chemical work to the point of being able to make a positive statement. I believe it is very likely that oxide and silicate were produced. The only thing I will absolutely vouch for is the elementary portion of the analysis, the silica being present to the extent named and the lead and zinc being there in the respective amounts given.

H. M. ST. JOHN: A number of years ago, three of us had a paper before the American Institute of Metals on somewhat this same subject,

* Detroit Lubricator Co., Detroit.

and, although working with quite a different alloy, Mr. Roast's results agree reasonably well with our own. There is one difference that I would like to point out, however, because it probably is of interest to practical foundrymen. Mr. Roast noticed that the effect of silicon in connection with these particular alloys is less pronounced on the structure than is the effect of the aluminum. That, I believe, is true in such a case as this, where the percentage of zinc is very low. With a higher percentage of zinc, the effect of silicon on the structure becomes more and more pronounced until, at somewhere around 10 per cent zinc, it is hard to tell the difference between silicon and aluminum, so far as the structure is concerned.

One other point I think should be remembered, namely, that the percentages of impurities given by Mr. Roast are those added. The amount of impurity left in the finished metal will depend on melting conditions, because all of these impurities are very readily oxidizable and more or less of them will be present in the finished metal, according to the degree of oxidation which can take place in the melting atmosphere.

Melting them as Mr. Roast did, in a crucible, well protected with charcoal, there is no very strong oxidizing condition. The condition is much more oxidizing in the case of the rocking-type electric furnace, in which our own experiments were carried out. Thus, by adding a small percentage of impurity, in our case we undoubtedly had more left in the finished metal than was the case with Mr. Roast's experiment, and the effects were a little more pronounced than those which have been described by the author.

G. H. CLAMER.* A number of years ago we were extensively engaged in that same line of manufacture, namely, railway bearings; we discontinued our foundry operations in that line seven or eight years ago, I believe. Prior to that time, the newer alloys containing silicon and aluminum had not yet appeared in the scrap to any considerable degree, and it was only very rarely that we would find a heat containing silicon or aluminum.

As Mr. Roast has pointed out, it is a very easy matter to identify these impurities by the appearance of the castings. About the time we quit that line, however, the scrap was becoming more and more contaminated and we had about reached the conclusion that if we were to continue in manufacturing that line, it would be necessary to do our melting in a hearth furnace having a distinctly oxidizing atmosphere, in that way very easily getting rid of these two readily oxidizable metals.

The presence of phosphorus was always troublesome, of course; there was more or less phosphor-bronze getting into the scrap. The trouble was particularly acute when we had combinations of zinc and phosphorus. The A.R.A. specifications and also the A.S.T.M. specifications of seven or eight years ago restricted the zinc content to such a degree that it was necessary, in order to meet the specifications, to use a fairly large percentage of new metal (I am referring now to engine castings) or to conduct a refining operation in order to reduce the zinc content below the maximum allowance in the specifications.

* Ajax Metal Co., Philadelphia.

That also pointed the way to the use of the reverberatory type of furnace, or a furnace in which that element could be reduced. However, it was overcome in another way, namely, by making the specifications less rigid so far as the zinc was concerned; a higher zinc content was permitted. This change in specifications resulted from a conference between the railway representatives and the manufacturers, which I organized some time after the war. Such zinc content is beneficial in the production of castings and it permits of a wider use of scrap.

I note that Mr. Roast used an alloy containing about 2 per cent zinc; that alloy is practically within the specifications to which I referred. In those days, when we had a heat showing aluminum or silicon, it was necessary to discard it. Of course, as Mr. Roast has stated, the appearance of castings from such heats was entirely different from those of heats that did not contain those elements, and they would be rejected by the railway inspector if not rejected by the manufacturer. It is a very simple matter indeed to get rid of them, and I do not believe that the bearing manufacturers today experience any trouble whatever from that source.

MEMBER: Has the author tried to machine any of those silvery-colored skin castings, and has he noticed any increase in hardness or difficulty of machining?

H. J. ROAST: We did not machine any of the castings. They would ordinarily have been machined and we probably would have had trouble. I think it is very unlikely that in the machining of them you would not get trouble with your tools, but we did not make that test. We were dealing with the question of surface appearance as passed upon by the inspector.

Mention has been made of the higher zinc, by Mr. St. John, and of the silicon giving more trouble—with which, of course, I quite agree.

In the matter of loss of deoxidation, there is, in the paper, reference to the fact of probable loss due to oxidation and, to my way of thinking, that rather enhances what little value there may be to the experiment, because at least we did not have more than 0.05 per cent and, in all probability, may easily have had only half of that amount. Let us say, therefore, for argument's sake, that 0.02 per cent was absolutely inimical to the final casting.

Mr. Clamer has spoken of the need of an oxidizing atmosphere and of the use of the reverberatory furnace. We had a similar experience in our own business, in that we were getting worse scrap and wanted to refine it, but we did not want to refine it at our own expense. There was very little leeway in the railway business, at best, and we made very distinct attempts to get the railway to enter into the spirit of refining and unifying all their metal, which I am still convinced is the best thing and which eventually some railways, at least, will have to do.

Our experience, unfortunately, was not the same as Mr. Clamer's. As I understand it, he was able to get a lowered specification with the deterioration of the scrap. Unfortunately, in Canada it followed the reverse process—as the scrap got worse, the specifications became more rigid, which was not so nice.

In regard to having no trouble, we still do—and when I say "we,"

I am speaking of more than one manufacturer in Canada; we still have trouble because of those two facts—on the one hand, that the scrap is not improving (one can hardly expect it to improve, with the new additions of alloys in scrap coming from various sources), and, on the other hand, that we are not getting a decrease, but rather, an increase in the requirements of the specifications. Apparently there is not sufficient margin from the commercial point of view to treat this metal on a reverberatory basis. There, again, possibly the condition is different from that in the United States because our quantities are smaller. It may be that you can refine your metal in the reverberatory furnace under the same price conditions as hold for the ordinary method. That is not true with us in Canada.

G. H. CLAMER: I would like to mention, in connection with Mr. Roast's statement, that I had no idea in mind of a double operation, namely, first of refining the metal and pigging it, and then remelting. The suggestion was to melt it directly in the reverberatory furnace and, if the metal was found to contain silicon and aluminum, to refine and then cast. As Mr. Roast has stated, the margin of profit is so small that it would not pay to conduct a double operation.

The evidence has been distinctly in favor of these wider specifications. Foundry troubles have been very largely reduced because of the higher zinc content. I think there is very good evidence that the service rendered by such bearings is at least equal to that of bearings made to the closer specifications.

H. J. ROAST: I am entirely in agreement with Mr. Clamer. It is only our inability up to date to get as far as has Mr. Clamer that prevents us from doing so.

CHAIRMAN SAM TOUR: We have recently had considerable experience in the use of silicon brass, as it is becoming more and more common, where we have considerably more silicon and zinc present than in the author's castings, and yet there is no indication of any such white powder or the formation of silicates of zinc.

G. H. CLAMER: I should not expect it to be a zinc silicate. So far as I know, we have never found any silicate. I should say that it is an oxide, entirely.

CHAIRMAN SAM TOUR: Of course, the old orthodox opinion was that it was all lead silicide. That may be all wrong, but I have always considered it; in the past years it has been rather commonly considered that it was lead silicide. Yet it would not appear as a light fluffy powder if it were lead silicide. What it is, I do not know.

In this particular connection, I might say that one convenient method of testing things of that nature is with the spectrograph. It will not tell whether this was lead silicide, possibly, unless one could pick out a particular crystal of it or a particular particle and test it spectrographically to find out the composition of that particular particle. I might also mention that the spectrograph might be a convenient method for determining these hundredths of a per cent of aluminum present, which is a rather mean determination normally in the laboratory.

* Lucius Pitkin, Inc., New York.

H. J. ROAST: I do not question for a moment that the white powder has not been seen in the methods and alloys used by the Chairman and Mr. Clamer; but I am absolutely sure, as you would be had you been in the foundry, that in the case of the castings referred to, under the conditions named, there is no question at all about the white appearance, none about the lightness of the fluffy powder, and none about the analysis of it containing silica.

I believe I can reasonably say that every precaution was taken to avoid getting in sand and silica from an outside source. I am quite open-minded on the subject of the composition of the silica—or the silicon, rather, which is its elemental form—and the other two elements; but for those of you who are in the foundry business and who think that they can carry silicon in such form without having the appearance as stated, I feel quite sure that the paper could be substantiated.

You will have noticed for yourselves the difference between the appearance of aluminum in the small casting, for instance, and in the large casting. Had we been dealing only with the large casting, we should have said: "No white effect." The statement must be taken strictly within the limits described.

H. M. ST. JOHN: In dealing with these alloys we always have a very complicated situation. Whether one gets zinc silicate or lead silicide, or in just what form the impurities are present, will depend a great deal not only on the nominal composition of the alloy but also on other impurities which may be present. We found, for example, that the effect of silicon was greatly influenced by the presence of tin—how, we did not know. We did know, however, that if we used a straight copper-zinc-lead, while the silicon had a bad effect, its effect and appearance were quite different than when we had copper, tin, zinc and lead. I am inclined to think, while I have no definite proof, that Mr. Roast is correct in assuming that zinc silicate is present in the case he mentions.

Effect of Superheat on Annealing of Malleable Iron*

By A. E. WHITE** AND R. SCHNEIDEWIND,† ANN ARBOR, MICH.

Abstract

As available data are meager on the effects of melting temperatures on the structure and annealing rates of white iron to produce malleable castings, the authors conducted an investigation on effects of superheat to see if some means would be revealed for appreciably reducing the time necessary to produce malleable iron. Three groups of irons were investigated, the first consisting of specimens of standard foundry composition, the second and third being made in a rocking-arc electric furnace of a low carbon and low silicon type. Superheat temperatures ranged up to 2970 Degs. Fahr. for the commercial irons, and up to 3180 Degs. Fahr. for the electric furnace irons. Times necessary to malleableize were noted. A conclusion reached was that superheating should be of very practical interest, since superheating along with good furnace design are capable of reducing normal malleableizing cycles up to 50 per cent.

1. The subject of malleabilization of white iron has received considerable attention in the last decade, and many attempts have been made to better understand the mechanism of graphitization and to develop a means of shortening the annealing cycle. In general, it can be stated that the annealing characteristics of an iron can be varied by changing one or more of the following factors:

* This paper is a portion of a thesis submitted in partial fulfillment of the requirements for the degree of Doctor of Philosophy in metallurgy at the University of Michigan.

** Professor of Metallurgical Engineering and Director of Department of Engineering Research, University of Michigan.

† Research Engineer, Department of Engineering Research, University of Michigan.

NOTE: This paper was presented and discussed before one of the malleable iron sessions at the 1933 Convention of the American Foundrymen's Association.

- (a) Raw materials.
- (b) Composition.
- (c) Melting and casting conditions
- (d) Annealing conditions.

2. To date, the subject of composition and the nature of the annealing cycle, furnace design and rate of cooling have been investigated rather extensively. The matter of raw materials is receiving serious attention in some of the large malleable iron foundries.

3. It has been pointed out that the temperature of melting affects the structure of gray cast and the annealing rates of white cast iron. The work of Tanimura¹ is outstanding on gray irons. His results show that by increasing the amount of superheat, but with a constant pouring temperature, the graphite particles decrease in size with accompanying increase in physical properties until a point is reached at which it is difficult to obtain purely pearlitic irons, since the graphite particles have a tendency to be surrounded by ferrite envelopes.

4. The data on white irons, on the other hand, are not so complete, being often confused with the pouring temperature, and in some cases it is contradictory. In most cases no quantitative conclusions can be drawn. It was felt that a study of the effects of superheat on the annealing characteristics of malleable iron might reveal a means of appreciably reducing the time necessary to produce malleable irons.

Description of Irons Studied

5. Three groups of irons were investigated. The first was the standard composition of a large foundry and was prepared under production conditions at that plant. The composition was nominally 2.75 per cent carbon and 1.15 per cent silicon. The second and third sets were made in a Detroit Electric rocking furnace of the low carbon and high silicon type. One set contained 1.4 per cent carbon and 1.5 per cent silicon, the other 1.5 per cent carbon and 1.7 per cent silicon. These were poured into standard test bars in sand molds.

Commercial Iron

6. A study was first made on commercial irons made at one of the large Midwestern malleable plants. The metal was that used in their regular production.

¹ Tanimura, H., "Influences of Cooling Velocity and Melting Temperature on the Graphitization of Cast Iron." *Memoirs, College of Engineering, Kyushu Imperial University, Fukuoka, Japan*, v. 6, no. 2, 1931.

Table 1
ANALYSES OF COMMERCIAL IRONS

Designation.	Superheat, degs. Fahr.	Carbon, per cent.	Silicon, per cent.	Manganese, per cent.	Sulphur, per cent.	Phosphorus, per cent.
S-1	2970	2.80	1.19	0.36	0.095	0.08
S-2	2900	2.71	1.17	0.35	0.095	0.08
S-3	2750	2.71	1.18	0.35	0.095	0.08

Table 2
MALLEARIALIZATION OF COMMERCIAL IRONS

Designation	S-1	S-2	S-3
Superheat, degs. Fahr.....	2970	2900	2750
Time to 1700 degs. Fahr., hrs.....	0.5	0.5	0.5
Time at 1700 degs. Fahr., hrs.....	20.0	25.5	31.0 1st Stage
Time 1700 to 1400 degs. Fahr., hrs.....	4.0	4.0	4.0
Time 1400 to 1325 degs. Fahr., hrs.....	1.0	1.0	1.0
Time at 1325 degs. Fahr., hrs.....	26.0	31.0	34.0 2nd Stage
Total	51.5	61.0	70.5

7. The usual procedure at this plant is to melt in the cupola at about 2700 degs. Fahr., and to refine and superheat in a Pittsburgh electric furnace at 2900 to 2925 degs. Fahr. In this investigation the process was so modified that for periods of time the production melts were superheated to 2750, 2900 and 2970 degs. Fahr., respectively. Temperatures in excess of 2970 degs. Fahr. could not be obtained at this plant because the refractories in the electric furnace were not suitable for the higher temperatures.

8. In all cases the test-bar molds were placed in the production line and the bars were cast under conditions identical with their regular work. Pouring temperatures were close to 2750 degs Fahr. The analyses of the irons are given in Table 1.

9. Annealing tests were conducted in a muffle type furnace the atmosphere of which was controlled by burning metered quantities of gas and air, in order to give a ratio of $\frac{CO_2}{CO} = 0.40$. This ratio was found to prevent scaling and decarburization, as was pointed out by the authors in a previous paper.²

10. The time necessary to complete first-stage graphitization, that is, to decompose completely all the massive cementite, was determined by heating specimens of these irons for various lengths of time at 1700 degs. Fahr. and examining metallographically.

² White, A. E., and Schneidewind, R., *The Metallurgy of Malleable Iron*. TRANSACTIONS, A.F.A. (1932), v. 40, pp. 88-116.

11. Second-stage graphitization was studied at 1325 degs. Fahr. Specimens were heated to equilibrium at 1700 degs. Fahr., cooled in 4 hours to 1400 degs. Fahr., then cooled to 1325 degs. Fahr. in one hour and held at this temperature for various lengths of time. The completeness of malleabilization was determined

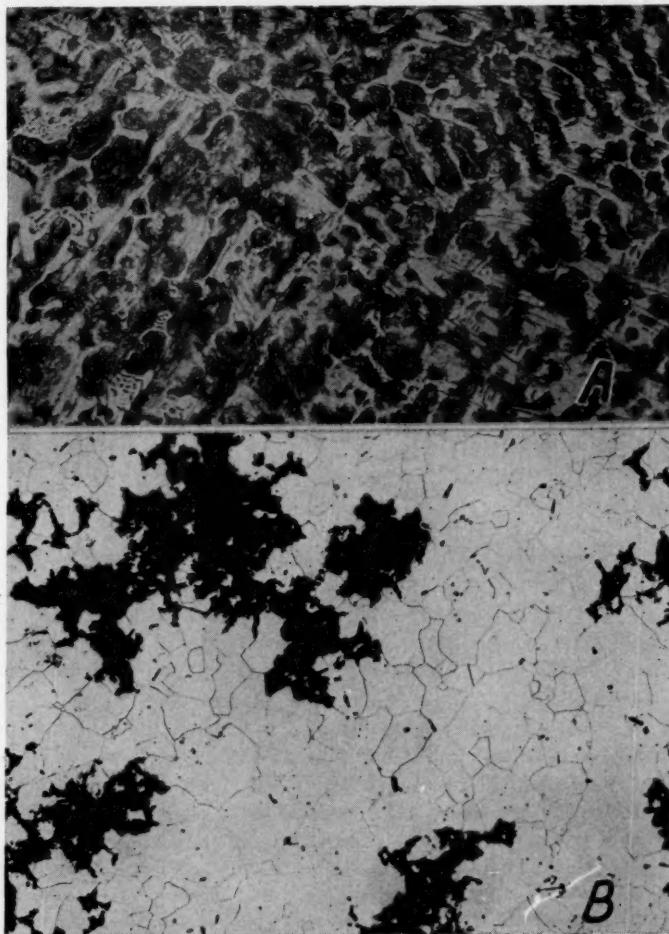


FIG. 1—PHOTOMICROGRAPHS OF IRON S-2. A: AS CAST; ETCHED IN 3% NITAL X100. THIS STRUCTURE IS REPRESENTATIVE OF THE THREE IRONS OF THIS GROUP. B: MALLEABLEIZED; ETCHED IN 3% NITAL X100. THIS STRUCTURE IS REPRESENTATIVE OF THE THREE IRONS OF THIS GROUP. GRAPHITE PARTICLES ARE LARGE, DUE TO HIGH CARBON CONTENT, 2.75 PER CENT.

Table 3
PHYSICAL PROPERTIES OF IRONS

Designation.	Superheat, degs. Fahr.	Tensile Strength, lbs. per sq. in.	Yield Point, lbs. per sq. in.	Elongation, per cent.
S-1	2970	49,450	46,150	14
		49,350	46,200	14
		49,900	46,400	14
S-2	2900	49,700	46,200	12.5
		49,200	46,000	12.5
		48,400	45,900	12.5
S-3	2750	48,900	40,600	12.5
		47,950	41,700	12.5
		50,000	38,800	18.75
		48,600	38,750	12.5

Table 4
MALLEABILIZATION OF ELECTRIC FURNACE IRONS

Designation	E-2	E-5
Superheat, degs. Fahr.....	3150	2865
Per cent Carbon.....	1.40	1.40
Per cent Silicon.....	1.56	1.54
Time to 1700 degs. Fahr., hrs.....	0.5	0.5
Time at 1700 degs. Fahr., hrs.....	3.0	3.5 1st Stage
Time 1700 to 1400 degs. Fahr., hrs.....	1.0	1.0
Time 1400 to 1325 degs. Fahr., hrs.....	1.0	1.0
Time at 1325 degs. Fahr., hrs.....	7.5	15.0 2nd Stage
 Total.....	12.0	20.0

metallographically. The superheat and time for complete malleabilization are given in Table 2.

12. It is evident that increasing the temperature from 2750 to 2900 degs. Fahr. brings about a shortening of the annealing cycle of from 10 to 12 per cent. Increasing the temperature to 2970 degs. Fahr. results in a saving of time of 28 per cent.

13. The photomicrograph of Fig. 1-A shows the structure of Iron S-2 in the "as cast" condition which is typical of all three irons. The photomicrograph of Fig. 1-B shows a structure representative of these irons after annealing. The graphite particles are quite large, due to the high carbon content.

14. Physical properties of these three irons were determined and are given in Table 3.

Electric Furnace Irons

15. In order to check these conclusions, two irons of a low-carbon, high-silicon type having a high strength were made at the

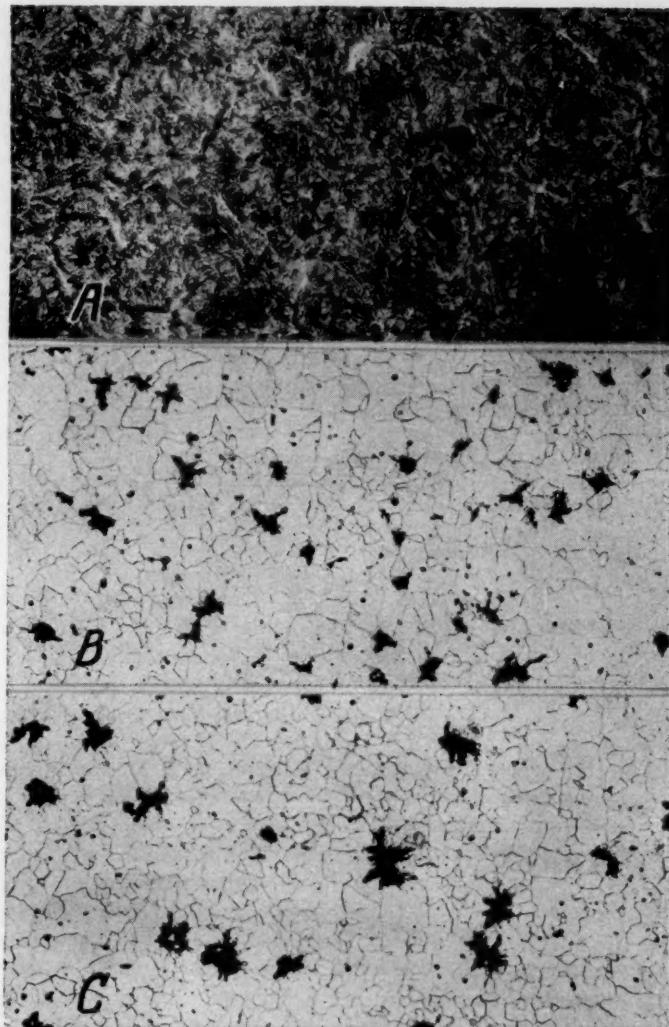


FIG. 2—PHOTOMICROGRAPHS OF IRON E-2 (A, B), AND IRON E-5 (C). A: AS CAST; ETCHED IN 3% NITAL. X100. THIS STRUCTURE TYPICAL OF THE 5 ELECTRIC FURNACE IRONS DESCRIBED IN PAPER. B: MALLEABLEIZED; ETCHED IN 3% NITAL. X100. C: MALLEABLEIZED; ETCHED IN 3% NITAL. X100. NOTE THAT GRAPHITE PARTICLES ARE LARGER THAN IN IRON E-2.

Table 5
PHYSICAL PROPERTIES OF IRONS

Designation.	Superheat, degs. Fahr.	Tensile Strength, lbs. per sq. in.	Yield Point, lbs. per sq. in.	Elongation, per cent.
E-2	3150	69,400	49,650	12.5
		65,400	50,500	9.0
		60,000	44,800	10.5
		64,200	44,600	17.0
E-2	2865	61,000	—	17.0
		61,750	58,000	18.75

Table 6
MALLEABILIZATION OF ELECTRIC FURNACE IRONS

Designation	E-1	E-7	E-6
Superheat, degs. Fahr.	3180	3030	2850
Per cent Carbon	1.39	1.51	1.50
Per cent Silicon	1.71	1.65	1.70
Time to 1700 degs. Fahr., hrs.	0.5	0.5	0.5
Time at 1700 degs. Fahr., hrs.	1.5	1.75	4.0 1st Stage
Time 1700 to 1400 degs. Fahr., hrs.	1.0	1.0	1.0
Time 1400 to 1325 degs. Fahr., hrs.	1.0	1.0	1.0
Time at 1325 degs. Fahr., hrs.	1.0	4.5	5.0 2nd Stage
Total	5.0	8.75	11.5

University of Michigan in a Detroit Electric Furnace. One of the irons was superheated at 3150, the other at 2865 degs. Fahr. The pouring temperature from the ladle was 2750 degs. Fahr. in both cases. Annealing characteristics were determined as described above and the results, along with the analyses, are given in Table 4.

16. In this set, the greatest acceleration in annealing time took place in the second stage. The total annealing cycle was shortened in the neighborhood of 40 per cent by increasing the temperature of the melt from 2865 to 3150 degs. Fahr.

17. The structure of Iron E-2 is shown in the photomicrograph of Fig. 2-*A* and is representative of all these electric furnace heats. Figs. 2-*B* and 2-*C* are photomicrographs which show the structure of Irons E-2 and E-5, respectively. Superheat favors the formation of smaller but more numerous graphite particles. The physical properties are shown in Table 5.

18. Another series of irons of the low-carbon, high-silicon, high-strength type was prepared. The superheat temperatures were 3180, 3030, and 2850 degs. Fahr., respectively. The annealing characteristics determined as above are given in Table 6.

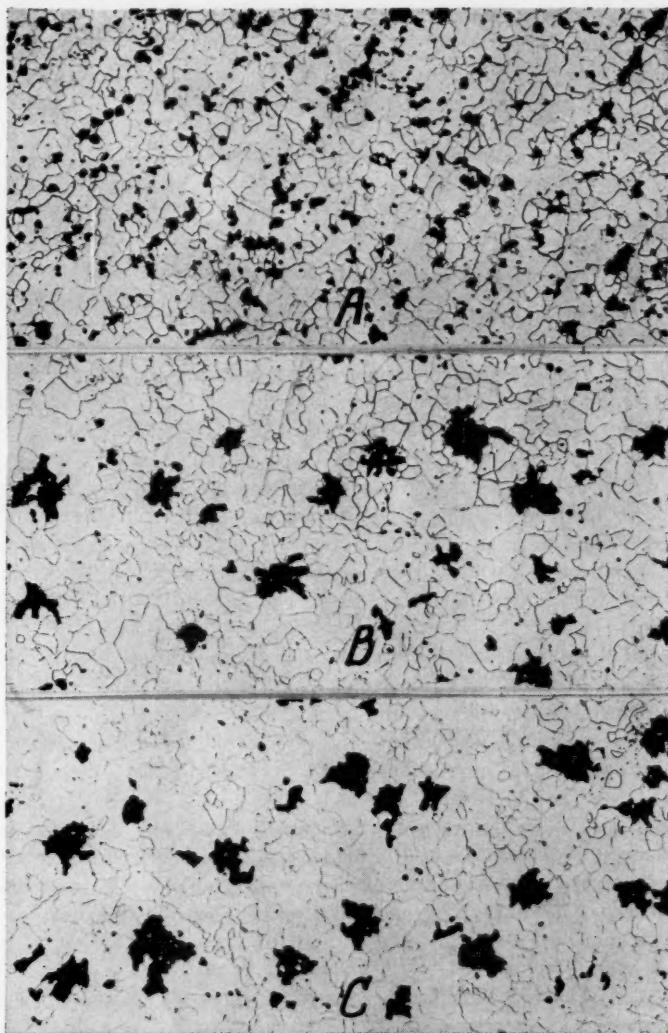


FIG. 3—PHOTOMICROGRAPHS OF IRON E-1 (A), IRON E-7 (B), AND IRON E-6 (C). A: MALLEABLEIZED; ETCHED IN 3% NITAL. X100. B: MALLEABLEIZED; ETCHED IN 3% NITAL. X100. C: MALLEABLEIZED; ETCHED IN 3% NITAL X100. NOTE THAT SUPERHEAT FAVORS THE FORMATION OF SMALLER GRAPHITE PARTICLES.

Table 7
PHYSICAL PROPERTIES OF IRONS

Designation.	Superheat, degs. Fahr.	Tensile Strength, lbs. per sq. in.	Yield Point, lbs. per sq. in.	Elongation, per cent.
E-1	3180	65,600	47,750	12.0
		66,750	49,000	10.0
		60,500	48,700	10.0
		59,700	50,200	9.5
		66,000	50,500	8.5*
E-7	3030	60,100	40,700	18.6
		63,150	44,400	18.75
		63,500	42,000	14.0*
		62,000	—	18.0
E-6	2850	All bars contained blow holes		

* Broke outside gage length.

19. Again the specimen having a high degree of superheat was found to malleableize in a much shorter time than irons of the same composition not heated to so high a temperature in the molten state.

20. One hour's time was taken with all irons in cooling from 1400 to 1325 degs. Fahr. The critical temperature lies near 1375 degs. Fahr. for this iron and near 1350 degs. Fahr. for irons of lower silicon contents. Some second-stage annealing is accomplished during the temperature drop from the critical to 1325 degs. Fahr.

21. In most cases the second stage graphitization done at this time is negligible, but in specimen E-1, due to the rapidity of the reaction, this one-hour cooling period accomplished at least half of the second stage annealing. Hence, a value of 2.5 hours is used in all calculations instead of 1 hour, in order to make the results comparable with the above irons.

22. The photomicrographs of Figs. 3-A, 3-B and 3-C show the structure of these irons at 100 diameters. Again increasing superheat favored the formation of more numerous, smaller graphite particles. The physical properties are as shown in Table 7.

Discussion of Results

23. The full significance of these data is not gained without correlating the results of all three series of irons. As a first step, the time necessary for first and second stage annealing must be added, for this sum is a measure of the time it would take in practice to manufacture malleable iron. This information is given in Table 8.

Table 8
EFFECT OF SUPERHEAT ON MALLEABLEIZING TIME

Designation.	Superheat, degs. Fahr.	Annealing Time, hours.
S-1	2970	46
S-2	2900	56.5
S-3	2750	62.5
Reference	2800	62.0
E-2	3150	10.5
E-5	2865	18.5
Reference	2800	19.5
E-1	3180	4.0*
E-7	3030	6.25
E-6	2850	9.0
Reference	2800	9.6

* Corrected value.

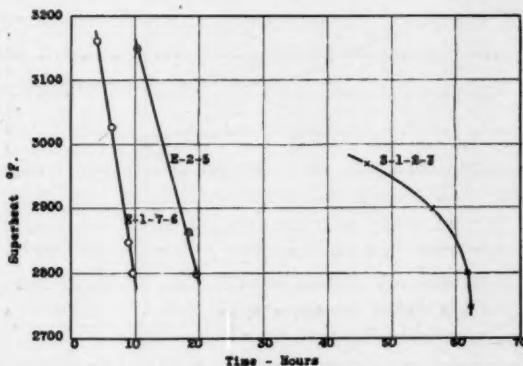


FIG. 4—GRAPH SHOWING EFFECT OF SUPERHEAT ON SECOND-STAGE ANNEALING TIME.

24. When these values are plotted, as in Fig. 4, the annealing time at 2800 degs. Fahr. can be obtained by interpolation or extrapolation. For the series beginning with S-1, this point is 62 hours; for the series beginning with E-2, it is 19.5 hours; for the series beginning with E-1, it is 9.6 hours. These values at 2800 degs. Fahr. will be used as reference values.

25. The beneficial effect of superheat can be seen by comparing the malleableizing time of the various irons with the value of its type at 2800 degs. Fahr. Table 9 shows this comparison on a per cent basis.

26. These values are presented graphically in Fig. 5. A study of the graph shows that acceleration in annealing rates becomes appreciable when melt temperatures are above 2900 degs.

EFFECT OF SUPERHEAT ON MALLEABLE ANNEALING

Table 9
COMPARISON OF SUPERHEATED IRONS

Designation	Superheat degs. Fahr.	Annealing Time Compared with that at 2800 degs. Fahr., per cent.
S-1	2970	74.3
S-2	2900	91
S-3	2750	101
Reference	2800	100
E-2	3150	53.8
E-5	2865	94.8
Reference	2800	100
E-1	3180	41.6
E-7	3030	65
E-6	2850	93.8
Reference	2800	100

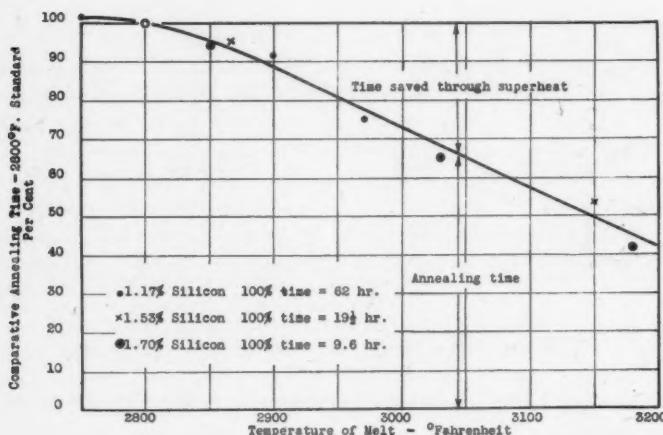


FIG. 5—GRAPH SHOWING EFFECT OF SUPERHEAT ON ANNEALING TIME.

Fahr. This curve indicates that very practical benefits may be obtained by superheating.

27. The matter of superheating should be of very practical interest, since superheat along with good furnace design are capable of reducing normal malleableizing cycles up to 50 per cent. A typical commercial cycle is as follows:

Time to bring to heat.....	35 hours
Soaking time	50 hours
Cooling to below critical.....	50 hours
Cooling to 1200.....	10 hours
 Total	 145 hours

28. It is the opinion of the writers that the use of a well-designed tunnel kiln can reduce the cycle at least 25 hours. The elimination of the packing material in such a furnace will permit bringing to heat in 15 to 20 hours. The temperature can also be dropped from the soaking temperature to the critical in less time in a tunnel kiln. This is advisable because, when the work has been graphitized to equilibrium in the first stage, an excessively slow rate of cooling to the critical serves no useful purpose.

29. Progressive plants having facilities for superheating can effect further reductions of from 15 to 20 hours in the actual annealing time in irons of normal composition. Iron S-2 is in this manner being annealed in large tonnages in a total time cycle of 51 hours with a top or soaking temperature of 1720 to 1730 degs. Fahr. A shorter cycle will increase, at the same time, the effective life of the boxes in which the castings are placed, an appreciable item in calculating annealing costs.

Acknowledgment

30. This study was sponsored by the Committee on Industrial Gas Research of the American Gas Association. Its progress was greatly aided by the encouragement and counsel of E. D. Milener of that organization. Thanks are due the Ford Motor Car Co. and the Malleable Iron Research Institute for the white irons used and to the Advisory Committee on Short-Cycle Malleabilizing for their valuable advice and assistance in laying out the program of research.

DISCUSSION

MEMBER: How did that low-carbon iron machine? That is, how did it compare with normal malleable in machining?

R. SCHNEIDEWIND: It machined satisfactorily. There was not much difference—no difference that the machinist could determine.

CHAIRMAN FRED L. WOLF: I imagine that you have a refractory problem when superheating, do you not?

R. SCHNEIDEWIND: The representatives of an electric furnace com-

* Ohio Brass Co., Mansfield, Ohio.

pany helped make the low-carbon high-silicon iron. They wanted to use a high temperature in order to show that they could go up to 3250 degrees without destroying the refractory.

MEMBER: Have the authors developed any theoretical explanation for the effect of superheating?

R. SCHNEIDEWIND: Yes, I believe the superheat brings about super-cooling. We have obtained a similar result by the addition of small quantities of foreign materials in the same way that the modification of aluminum-silicon alloys is obtained. My theory of the modification and of superheat is that it brings about super-cooling. The result is that the freezing point is lower and the lower critical point is higher. By inducing super-cooling, one also obtains smaller crystals and a greater degree of meta-stability in the solid solution. Upon subsequent annealing, the carbon has an increased tendency to graphitize. I do not know how to explain the increased number of graphite nuclei obtained by superheat.

The speed with which a carbon atom travels through iron to a graphite nucleus at any given temperature does not change with increasing content of silicon or other material, or by superheat of the melt. On the other hand, superheat increases the number of graphite nuclei and, hence, reduces the distance between nuclei. Therefore, the distance a carbon atom must travel in order to graphitize becomes less and the time for malleableizing naturally becomes less.

MEMBER: You had actual evidence of super-cooling with an increase in the freezing point?

R. SCHNEIDEWIND: No. On gray iron, Tanimura obtained a decrease of 20 degs. Fahr. from the freezing point, and an increase of from 15 to 25 degs. Fahr. on the lower critical. We checked him on the lower critical on white irons. I merely used his theory and applied it to malleable.

MEMBER: Our experience has been that as the silicon goes higher, these nodules get smaller. We have a series of samples showing that very clearly. Consequently, I do not think it is quite right to attribute this condition to superheat and ignore the effect of silicon, because one might get the same thing with silicon as with superheat.

R. SCHNEIDEWIND: In a paper² I presented before this Association last year, I showed that with an increase of silicon there is an increased number of graphite nuclei. At 1832 degs. Fahr. there is a phase change in the iron-silicon diagram. The FeSi changes to Fe₃Si. With the increase in either silicon content or cooling rate or superheat, more of the FeSi is retained. It is true that increased silicon will produce smaller and more numerous graphite particles. However, by keeping the silicon constant and increasing the superheat, smaller and more numerous graphite particles likewise will be obtained.

MEMBER: What casting difficulties were encountered with reference to shrinks or cracks?

R. SCHNEIDEWIND: With regard to the first iron we described, that is made commercially and there are no particular difficulties. In fact, the manufacturers claim that they have less difficulties than do other

² White, A. E. and Schneidewind, R., *The Metallurgy of Malleabilization*. TRANSACTIONS A.F.A. (1932), v. 40, p. 88.

foundries. The lower carbon, high-silicon irons I do not believe are so good. It is possible, according to one gentleman here, to produce commercially irons having silicon up to 1.6 and 1.9 per cent. We encountered quite a bit of difficulty at the University in making the low-carbon high-silicon irons.

MEMBER: Do you believe, then, that we in the practical end of the foundry industry would get better iron if we melted and poured or tapped at high temperature, say, 1.1 per cent silicon iron?

R. SCHNEIDEWIND: I believe that if you could get 100 or 200 degs. Fahr. more temperature before tapping, you would get a faster annealing iron and, from the results I have obtained, the physical properties would not be impaired.

MEMBER: Would they be improved any?

R. SCHNEIDEWIND: I do not think there would be enough difference to notice. I have checked on physical property figures with the foundry that made these irons, and the agreement was satisfactory.

High-Strength and Wear-Resistant Malleable Cast Iron

BY REBECCA HALL,* CHICAGO

Abstract

In an endeavor to develop a wear-resistant malleable iron, the author experimented with the effect of chromium additions plus silicon control. It was particularly desired to secure castings with high strength, moderate ductility and considerable resistance to non-lubricated abrasion. The test castings were made by ladle additions to standard metal, the castings being given the regular anneal. Chromium additions ranged from 0.25 to 1.15 per cent, with the silicon varying from 1.02 to 2.02 per cent. Maximum tensile strength was secured with 0.84 per cent chromium and 1.57 per cent silicon. Elongations were decreased as the chromium was increased. Hardness also increased. Elastic limits increased up to 62,500 lbs. per sq. in. The results indicate possibilities for further research.

INTRODUCTION

1. In view of the progress made in the use of alloys in gray iron and steel during the past decade, it is not remarkable that there has been a great increase in the production of alloyed malleable iron in the past year. Up to that time, although the malleable industry had greatly improved its product through organized control, it was offering only one or two types of castings, as compared with many specialized and highly diversified types of gray iron and steel.

2. Apparently the malleable foundryman in the past has had a prejudice against alloy additions because most alloys either inhibit or promote graphitization, in either case affecting the annealing process. An article presented before this association in

* Metallurgist.

NOTE: This paper was presented and discussed before one of the malleable iron sessions at the 1933 Convention of American Foundrymen's Association.

1925[†] gave some of the first comprehensive data on the effects of alloy additions in malleable iron.

3. In the past, standard malleable has met every requirement for reliability and general mechanical properties. Today, however, like every other engineering material, it is being called upon to sustain greater loads and give better service than was formerly necessary. Besides demanding higher strength properties, customers ask for highly specialized characteristics.

4. For instance, the new so-called "railroad" specification calls for a yield point of 35,000 lbs. per sq. in. Until comparatively recently, the yield point was not considered when writing malleable specifications. Other qualities such as extreme ease of machining, resistance to heat, exceedingly high strength, and resistance to corrosion and wear, are also in demand.

5. With these facts in mind, the author has tried to develop a wear-resistant malleable iron, particularly to produce castings with high strength, moderate ductility and considerable resistance to non-lubricated abrasion. Before beginning experimental work, a complete bibliography and detailed study of the literature on alloyed malleable iron was made.

6. It was essential that this metal be of moderate cost and made with the least departure from ordinary malleable practice in the foundry and annealing room. Although all the care and accuracy possible were used, the values obtained are not research laboratory values. These tests were made in a commercial malleable foundry, with the usual procedure, under exactly the same conditions which prevail in production work; for this reason, they are practical enough to point the way toward a solution of this particular problem.

Reason for Adding Chromium and Silicon

7. Increased strength and hardness are necessary in a wear-resistant iron. In irons of the same base composition, increasing hardness indicates decrease of soft ferrite and increase of pearlite, with correspondingly higher wear resistance. It is well known that the relationship between Brinell hardness and wear resistance in many cases is controversial. However, for the purposes of the present investigation, Brinell hardness figures may be fairly considered as an accurate measure of wear resistance, and the author has therefore used Brinell figures throughout the investigation.

8. Chromium was added to raise the hardness and strength,

[†] Reference numbers as shown herein correspond with Bibliography at end of paper.

and silicon additions were made to counteract the inhibiting effect of chromium on the annealing process. For years, chromium has been taboo so far as the malleable foundry is concerned, because of this effect.^{2,3,4,5} Without additions of a graphitizing element such as nickel or silicon, chromium is very dangerous in amounts over 0.07 per cent, but silicon will counteract the effect of chromium since it increases softness and favors formation of temper carbon. Therefore, it seemed logical that by adding both chromium and silicon,⁶ the advantages of both and the disadvantages of neither could be obtained, as is accomplished in chromium-nickel steel, and a hard malleable iron with high strength and good ductility would result.

Experimental Work

9. The experimental work was kept as practical as possible. Chromium was added in amounts from 0.25 to 1.15 per cent with enough silicon to counteract its tendency to form stable carbides which might prevent the annealing operation. Earlier work indicated in a general way the amount of silicon necessary for this purpose, but two different percentages of silicon were added for each percentage of chromium to give more data on this ratio. All the foundry operations were carried out under accurate technical control.

10. The alloys used were ferrochrome (64.55 per cent chromium), ferrosilicon (90 per cent silicon), and in one case, zirconium-manganese-silicon (61.85 per cent silicon, 10.54 per cent

Table 1

Test Bar No.	Tensile Strength, lbs. per sq. in.	Elastic Limit, lbs. per sq. in.	Elongation in 2 in., %	Brinell Hardness	Cr %	Si %
3211*	54,500	37,500	18.5	131	0	1.02
C-22-A†	70,000	42,400	11.0	175	0.22	1.10
C-22-B†	67,700	41,000	12.5	175	0.22	1.10
C-23	72,900	43,600	9.0	179	0.23	1.24
C-24-A	70,000	43,500	7.0	168	0.24	1.27
C-24-B	75,600	47,100	9.0	168	0.24	1.27
C-24-C	70,400	42,800	9.5	168	0.24	1.27
C-37	76,900	49,250	9.5	179	0.37	1.28
C-40	68,500	48,000	6.0	183	0.40	1.44
C-56	80,500	52,500	6.0	201	0.56	1.39
C-73	67,700	54,750	4.5	212	0.73	1.66
C-84	80,000	59,000	4.0	213	0.84	1.57
C-87	63,500	52,500	3.0	201	0.87	1.62
C-108	76,800	62,000	3.0	223	1.08	1.72
C-114	67,000	62,500	2.5	202	1.14	2.02

*Standard unalloyed malleable with total carbon 2.33, manganese 0.37, phosphorus 0.16, and sulphur 0.10.

†Silicon added in the form of zirconium-manganese-silicon.

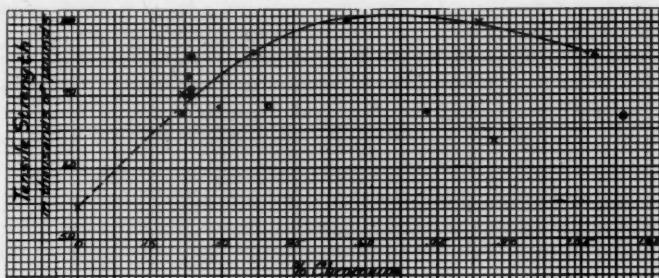


FIG. 1—EFFECT OF INCREASING CHROMIUM ON TENSILE STRENGTH. X : IRONS WHERE RATIO OF ADDITIONAL SILICON TO CHROMIUM IS LESS THAN 0.71; θ : WHERE THE RATIO LIES BETWEEN 0.71 AND 1.05.

manganese, 21.78 per cent zirconium). The special low-melting alloys, ground to 20 mesh and down, were used to insure complete solution in the ladle.

11. Additions of alloys were made in 50-lb. ladles. The metal was poured back and forth twice to obtain thorough mixing, and the bottom of the ladle was carefully inspected to make sure that none of the alloy remained undissolved. Recovery was about 97 per cent for chromium and 95 per cent for silicon.

12. Test bars were poured in the standard malleable size, $\frac{5}{8}$ in., and were given a standard 120-hour anneal at the usual temperature. They were pulled on a 100,000-lb. tensile machine, the yield point being measured by a precise extensometer.

13. The general results are shown both in tabular form (Table 1) and in graphical representation (Fig. 1).

TEST RESULTS

Tensile Strength.

14. The tensile strength of chromium-silicon malleable iron is from 10,000 to 25,000 lbs. per sq. in. more than that of ordinary malleable iron, which is generally given as 55,000 lbs. per sq. in. An effort is made to determine the effect of the ratio of the additional silicon to chromium, on the tensile strength. By additional silicon is meant that in excess of the normal amount in the base mixture; in other words, the excess silicon added in the ladle with the chromium.

15. In Fig. 1, X is used for irons where the ratio of additional silicon to chromium is less than 0.71, and θ where this value lies between 0.71 and 1.05. It is evident from Fig. 1 that the

lower ratio gives greater strength, as would have been expected from the individual effects of silicon and chromium on malleable iron.

Elastic Limit.

16. Increasing the chromium increases the yield point regardless of the amount of silicon present, as is shown in Fig. 2. It is interesting to note that the elastic limit of the chromium-silicon alloyed irons is not only higher in pounds per square inch but also is higher in per cent of ultimate strength than is that of ordinary malleable iron or cast steel. The A.S.T.M. specification A47-30 calls for yield point of 32,500 lbs. per sq. in., while the lowest one obtained on a chromium-silicon iron is 41,000 lbs. This great increase in the elastic limit is a matter of great commercial importance, for it has been recognized in recent years that the yield point, or point at which permanent deformation takes place, is often more significant than the ultimate breaking load.

Elongation.

17. Regardless of higher silicon, high ductilities are not maintained with high chromium present; but the author is of the opinion that the data obtained are insufficient to draw any conclusions on this point. In fact, it seems highly probable that a

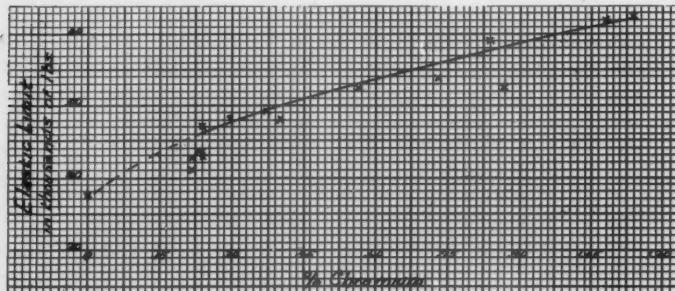


FIG. 2—EFFECT OF INCREASING CHROMIUM ON ELASTIC LIMIT.

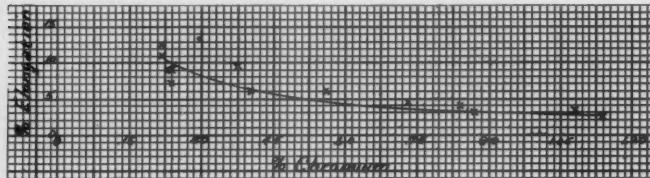


FIG. 3—EFFECT OF INCREASING CHROMIUM ON ELONGATION.

different heat treatment might increase the ductility materially. The effect of raised chromium on the elongation is shown in Fig. 3.

Hardness.

18. As shown in Fig. 4, the hardness as measured by Brinell hardness rises with increased chromium, except in two cases where so much silicon is present that it partially counteracts the hardening effect of chromium. This graph emphasizes the necessity of keeping the ratio of excess silicon to chromium in a technically regulated proportion.

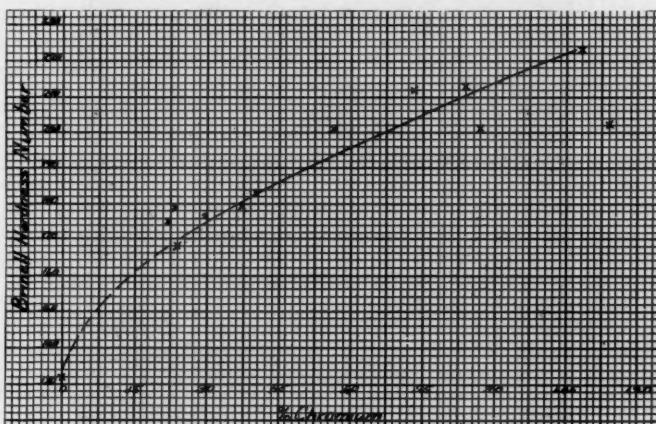


FIG. 4—EFFECT OF INCREASING CHROMIUM ON BRINELL HARDNESS.

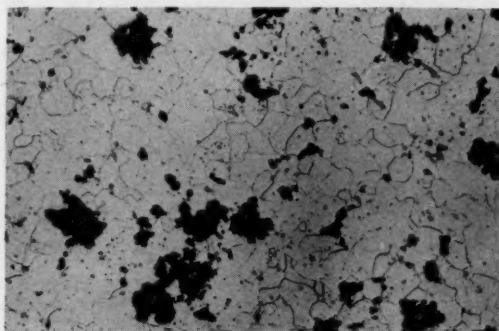


FIG. 5—MICROSTRUCTURE OF THE STANDARD MALLEABLE IRON TEST BAR. ETCHED WITH 4% NITAL; X100. TENSILE STRENGTH, 54,500 LBS. PER SQ. IN.; ELASTIC LIMIT, 37,500 LBS. PER SQ. IN.; ELONGATION IN 2 INCHES, 18.5 PER CENT; BRINELL HARDNESS, 131.

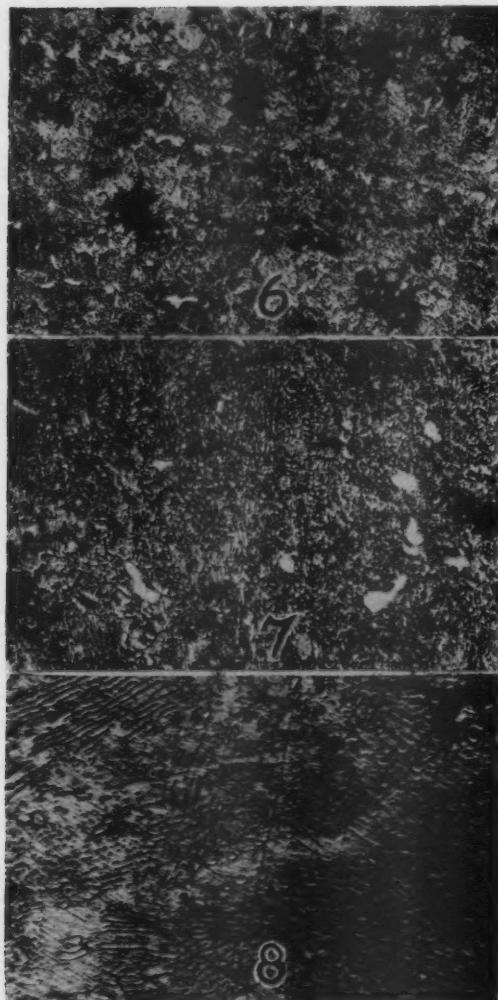


FIG. 6—STRUCTURE OF MALLEABLE IRON WITH 0.22 PER CENT CHROMIUM AND 1.10 PER CENT SILICON. ETCHED; X100. TENSILE STRENGTH 67,700 LBS. PER SQ. IN.; ELASTIC LIMIT 41,000 LBS. PER SQ. IN.; ELONGATION IN 2 IN., 12.5 PER CENT; BRINELL HARDNESS 175.

FIG. 7—STRUCTURE OF SAME IRON AS FIG. 6, BUT X375.

FIG. 8—STRUCTURE OF SAME IRON AS FIG. 6, BUT X800.

Ratio of Excess Silicon to Chromium.

19. In general, this ratio is not nearly so important as the absolute amount of chromium present. However, it explains some seeming discrepancies in results and should be considered when planning additions of this kind.

Microstructure

20. Fig. 5 illustrates the microstructure of the standard malleable iron test bar. The structure is temper carbon in the conventional form, surrounded by grains of ferrite, no pearlite being visible.

21. This iron is the best type of ductile malleable casting, meeting the A.S.T.M. Specification A-47-32T, grade 35018. The ferritic structure, of course, accounts for the softness and ductility of this standard malleable.

22. Fig. 6 shows, at 100 diameters, the excellent structure obtained by the use of small amounts of chromium with increased silicon. The temper carbon is slightly smaller than in the unalloyed malleable, and appears to be farther apart. A few patches of ferrite surround the temper carbon, and some very fine particles of unbroken carbides can be seen. The matrix is exceedingly finely laminated pearlite.

23. Fig. 7 shows the same structure at 375 diameters. A network of ferrite is visible, but even at this magnification the extremely fine structure is evident.

24. Fig. 8 shows the same sample at 800 diameters. The ground-mass, which at 100 diameters has the general appearance of a sorbitic pearlite, now can be seen to be almost entirely a very finely divided lamellar pearlite, surrounded by a network of ferrite. There is some evidence of a slightly spheroidized form in some of the pearlite, this undoubtedly being due to the well-known tendency of chromium-bearing pearlite to spheroidize at approximately 1550 degs. Fahr., which happened to be the annealing temperature in these experiments.

25. The finely laminated pearlite present in the matrix accounts for the hardness and high yield point of this iron and should contribute greatly to the wear resistance.

26. Higher chromium reduces the size of the temper carbon particles even more, as appears in Fig. 9. The matrix is composed of very finely laminated pearlite, as before, and more unbroken carbides are present in dendritic form. Physical properties are as follows:

Tensile strength, lbs. per sq. in.	80,500
Elastic limit, lbs. per sq. in.	52,500
Elongation in 2 in., per cent.	6.0
Brinell hardness.	201

27. The fine graphite and pearlite account for the great tensile strength of this bar, but the decreased elongation and increased hardness probably are due to the unbroken carbides.

28. Fig. 10 shows the exceedingly fine structure obtained with 1.14 per cent chromium and 2.02 per cent silicon. Very small graphite, very finely laminated pearlite, and more little patches of unbroken carbides can be seen. There is hardly any ferrite visible. Physical properties are as follows:

Tensile strength, lbs. per sq. in.	67,000
Elastic limit, lbs. per sq. in.	62,500
Elongation in 2 in., per cent.	2.5
Brinell hardness.	202

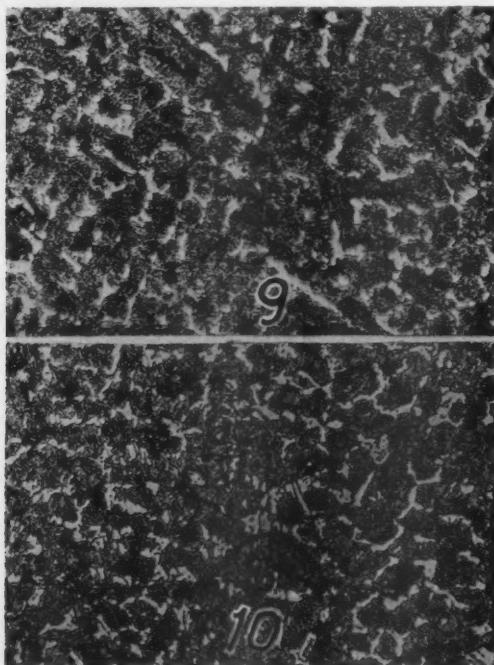


FIG. 9—STRUCTURE OF IRON WITH 0.56 PER CENT CHROMIUM AND 1.39 PER CENT SILICON. ETCHED; X100.

FIG. 10—STRUCTURE OF IRON WITH 1.14 PER CENT CHROMIUM AND 2.02 PER CENT SILICON. ETCHED; X100.



FIG. 11.—BENT TEST BAR OF IRON WITH 0.22 PER CENT CHROMIUM AND 1.10 PER CENT SILICON.

29. It is interesting to note that the elastic limit increases steadily with higher chromium, although so much silicon is present in this case that the strength is not so great as in other samples. The very low ductility and great hardness probably are caused by the unbroken carbides.

30. To summarize, additions of chromium and silicon in the ladle to standard malleable base mixtures change the structure of the annealed malleable so as to decrease the ferrite, refine the temper carbon, and cause a pearlitic structure. The effect of these changes on the physical properties is clearly understood when one remembers that ferrite has a tensile strength of about 50,000 lbs. per sq. in., and pearlite about 125,000 lbs. per sq. in., with 10 per cent elongation.⁷

Possibilities

31. This paper is offered in the nature of a progress report, and it is the hope of the author that it may open new fields of investigation. Possible uses for this type of metal will occur to any malleable executive.

32. No variations were made from the standard heat treatment, although the author recognizes the fact that very much greater ductility and other physical properties can be obtained by heat treatment specially designed for the solution of the stable carbides in such irons. However, since this series of tests was to be under strictly commercial production conditions, the regular annealing practice was followed. It will be interesting to

watch the development of this type of iron when special heat treatments are used.

33. These results also brought up the question of the use of chromium with higher silicon for the prevention of primary graphitization in heavy sections. The author is continuing this work at present.

34. It may be thought that elongations of from 7 to 13 per cent are not very high, but it must be remembered that within the last two decades very serviceable malleable castings were made to specifications requiring far less ductility and strength than the irons which form the basis of this paper. A brief comparison of physical properties of several irons illustrates this point, as follows:

	A.S.T.M. Cr-Si Spec.	A.S.T.M. Spec. 1910
Malleable 32510, 1930		
Tensile strength, lbs. per sq. in.	65,000	50,000
Yield point, lbs. per sq. in.	40,000	32,500
Elongation in 2 in., per cent ...	7.5	10.0
		2.5

35. The ductility of chromium-silicon malleable irons is shown in Fig. 11, which illustrates the bend obtained from one of the samples of No. C-22.

36. It is the belief of the author that this type of malleable casting will find many applications in commercial work. The combination of high strength, unusually high elastic limit, reasonable ductility and a hard wear-resistant matrix, all at very moderate cost, offers a metal which can be of great service in many fields.

37. The author wishes to express her appreciation to the Chicago Malleable Castings Co., at whose plant the bars were made and the chemical analyses and physical tests run; to the Electro-Metallurgical Co. for their kind cooperation; and to the Robt. W. Hunt Co. for the photomicrographs at 100 diameter.

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DISCUSSION

WRITTEN DISCUSSION

H. A. SCHWARTZ:* It has been known for some time that there are certain elements of which manganese and chromium are the best known, which retard graphitization considerably and do so more actively at temperature ranges above the critical point where the destruction of massive carbides is involved, than at temperatures below the critical point where pearlite is being graphitized.

It has been pointed out by certain workers at Iowa State College that these elements are those which enter into the carbide phase, and it is thermodynamically obvious that the vapor pressure of carbon from a carbide should be reduced if that carbide can dissolve or combine with other elements such as are here involved.

Miss Hall has chosen for her field of work the use of chromium for this purpose, somewhat modified in the interest of rapid first-stage graphitization by the addition of silicon. The physical results obtained will prove an interesting addition to the collection of information now available with regard to the properties of these pearlitic products.

The commentator would offer the suggestion that the results obtained should not be considered as the effect of chromium on the metal, but rather as the effect of combined carbon.

We are now confronted with a considerable range of commercial materials differing from malleable cast iron in that, instead of being ferritic, the metallic matrix is more or less pearlitic. This pearlitic character can be and has been obtained in many ways, (a) by suitable heat treatment of ordinary white iron, so conducting the anneal that pearlite remains behind in greater or less extent; (b) by the re-heat

* National Malleable & Steel Castings Co., Cleveland.

treating of ordinary malleable so that pearlite or sorbite are reformed and distributed in the manner and extent desired; and (c) by the addition of retarding elements with the intent that the normal annealing cycle shall produce that degree of pearlitic or sorbitic matrix desired. Miss Hall's material, of course, is in the latter class.

Now, it is probably not to be disputed that a pearlitic steel, containing at any rate the higher chromiums contemplated by the author, will depart in properties somewhat from a plain carbon steel; to that extent, chromium affects the particular alloys under consideration. However, their properties as differentiated from orthodox malleable cast iron are much more largely determined by the amount of pearlite present as an indirect consequence of the chromium than by the chromium *per se*. It is suspected, therefore, that the graphs shown in Figs 1, 2 and 3 are much more largely functions of the combined carbon content than of the chromium content, and that a plot of that character would have been quite useful in clearly understanding the significance of Miss Hall's experiments.

Whether materials whose matrix is not substantially pure ferrite should be called malleable cast iron, is something of a moot question. To the extent that they represent, at least moderately, ductile properties made by the graphitization of white cast irons, such a phrase is perhaps justified. To the extent that traditionally malleable iron has certain properties such as high elongation, great ease of machining, low hardness and high magnetic permeability, it would perhaps be better to coin a new name for these materials.

Such phrases as "incompletely" or "short annealed" malleable are open to misconstruction—the former, as carrying a connotation of inferiority; the latter, as a possible source of confusion with quick-annealed malleable. High-strength malleable has also this objection, since there should be some means of differentiating between ferritic malleable, reasonably strong because of low carbon content, and pearlitic products which are still stronger for the reason that Miss Hall's alloys have this property.

It would seem appropriate that the A.F.A. should take the lead in evolving a suitable nomenclature in this field, and this might much better be done promptly than later when perhaps an unfavorable terminology may have taken root.

ORAL DISCUSSION

FRANK M. ROBBINS: * I would like to endorse what Dr. Schwartz has said about the need for defining these pearlitic materials. The paper is very interesting and well worthy of study, but I question whether it should be presented at a malleable iron session. It seems to me that it is time for the A.F.A. to provide for the study of these pearlitic irons as a separate cast material.

* Ross-Meehan Foundries, Chattanooga, Tenn.

CHAIRMAN FRED L. WOLF:^{*} Has there been any practical application or specific application of this type of iron?

E. K. SMITH:[†] It is in service in several different commercial installations, and it was tried out quite carefully before it went into service. I would say that in six or eight months some definite data on service results will be available.

MEMBER: Have any impact test data been secured on this material?

E. K. SMITH: Impact tests are being made now. They are not yet completed, so they are not ready to report. Incidentally, since the paper was written, some work has been done on the corrosion of this material at the University of Wisconsin. A long series of different tests and different media are being run just to obtain the corrosion of this material as compared with that of standard malleable. Experiments on special heat treatments also are being run.

R. S. ARCHER:^{**} I wish to ask a question reflecting more the attitude of the user of malleable than that of the producer. I had the pleasure of seeing some of Miss Hall's work before the publication of this paper, and also some of the specimens, and it seemed to me that the bend tests were particularly good for pearlite malleable produced by the simple annealing process. I would like to hear some comments from some who may have done a great deal of work on this subject, as to whether that is the case. It is well known that very ductile products, having a pearlite or spheroidized carbide matrix, have been made by special annealing processes; but the specimens I had previously seen, which were made by the simple standard annealing practice, did not seem to give particularly good bend tests.

E. K. SMITH: Miss Hall showed me a lot of the samples she had and, in general, she got good bends with the lower chromes. As soon as the chrome got up a little higher and some cementite was left, the bends were not so good. In general, with 0.22 to 0.25 chrome, excellent bends were obtained.

R. S. ARCHER: That is not quite the question I had in mind. Products of a similar structure have been made by air cooling with standard composition. Such specimens have come to my attention, but they did not bend as well as this material under discussion. I was wondering if that was the general experience. Could similar bend tests be obtained with a Brinell hardness of 175, using the ordinary malleable composition?

E. K. SMITH: I can give only the results which the author obtained with the chrome.

MEMBER: You would not have a 175 Brinell hardness.

R. S. ARCHER: I mean with air cooling from above the critical.

MEMBER: Was that chrome addition one of 0.25 per cent or 0.025 per cent?

E. K. SMITH: The addition was 0.25 per cent, with the additional silicon.

MEMBER: I would like to ask a question with regard to the tests for

^{*} Ohio Brass Co., Mansfield, Ohio.

[†] Electro Metallurgical Co., Pittsburgh.

^{**} A. O. Smith Corp., Milwaukee.

wear resistance. How were those tests conducted, and with what other material were the tests compared?

E. K. SMITH: There were no tests made on wear resistance. This particular work was done for a particular job where the information was sufficient to know that if they could get a pearlitic structure with a certain definite Brinell, it probably would meet the requirements. As the equipment would be very expensive and it would take a long time to make the tests, the author did not attempt any wear tests.

WRITTEN REPLY BY AUTHOR

REBECCA HALL: I agree with Mr. Schwartz that the properties obtained in chromium-silicon malleable iron are partly the indirect result of the chromium present; but I believe that it is also partly the direct result of the chromium present dissolved in the ferrite. In any case, since the net result is due either to the direct effect of chromium on the ferrite or to the indirect effect of chromium in producing pearlite, it seems proper to plot physical properties as against chromium content.

Sand Control on a Continuous Molding Unit in a Malleable Foundry

BY CHARLES MORRISON,* SAGINAW, MICH.

Abstract

In foundries that have continuous molding and sand handling equipment, the use of facing is undesirable as it hampers production. There is a decided tendency toward use of 100 per cent facing sand. This paper is devoted to the description of sand control as practiced in a malleable iron foundry having continuous molding and sand handling equipment and which uses all facing sand.

Raw Material Storage

1. All raw materials are stored in the sand building. Unloading from the cars is done with a bucket and a 10-ton traveling crane. There are bins for storage of bank sand, sharp sand, facing sand, burnt sand, blast sand, black sand, clay and coal. For immediate use, small quantities of the above materials are stored in hoppers, which are filled by the crane. There are also hoppers for the storage of seacoal, molding sand and core mixtures. All hoppers are serviced by a traveling weigh lorry, the various hoppers being covered with tarpaulin to prevent contamination.

2. All molding and core sand mixtures are made in the sand storage building. The ingredients that go into the batches are weighed by the weigh lorry and deposited into the mixers and mullers. Batches of molding sand mixtures are weighed on a platform scale before entering the foundry, in order that accurate records of additions to the sand system can be kept.

* Assistant Metallurgist, Saginaw Malleable Iron Division, General Motors Corp.

NOTE: This paper was presented and discussed before one of the malleable iron sessions at the 1933 Convention of American Foundrymen's Association.

Sand System.

3. The conditioned sand is stored in two 100-ton tanks. The feed from the tanks is alternated so that one tank is discharging while the other is filling, thus allowing the sand ample time to cool and temper thoroughly. After leaving the tank the sand passes to the elevator, from which it is discharged into the revivifier. From the revivifier the sand is dropped onto a belt which carries it to the flight conveyors. The molding hoppers are located beneath the flight conveyors and the sand spills into them as it travels along, the surplus falling into the tail spouts, from where it is conveyed to the shakeouts.

4. The molds are dumped from the molding conveyor onto vibrating screens which remove the sand from the castings, and the latter then are loaded into buckets and carried to the knock-out conveyor, where the gates are broken off and the castings sorted into boxes. The sprue and sand are discharged into a mill, the burnt sand is taken off and screened, and the clean sprue is discharged into a box.

5. The reciprocators are located beneath the shakeout screens and rake the sand onto the shakeout belt. The rebonding mixture feeds onto the tail pulley of this belt. The sand is then carried up to the magnetic pulley and falls on the vibrating screens, where pellets of iron, core particles, clay balls, etc., are removed.

6. From the screens, the sand is discharged into the pug mill. As it falls into the pug mill, water is added through a series of jets. The sea coal feeder is located at this point and spills sea coal mixture into the pug mill. The reconditioned sand then is elevated and dropped into the storage tank, where it is allowed to cool and temper before use.

Laboratory Control

7. The sand laboratory supervises sand storage, core sand mixtures, molding sand and research. Tests for moisture, permeability and bond are taken on the molding sand every hour, or more often if necessary. Periodic tests are made on raw materials, cores, core batches and molding sand. Research is conducted at all times.

8. The man in charge of sand must cooperate closely with the core and molding departments in order to obtain the proper control.

9. Although laboratory tests show the sand and cores to be

within the specified limits of control, this does not necessarily imply that they are good cores and good sand. Many conditions beyond the bounds of test, influence the quality of these materials. Intimate knowledge of every condition pertaining to the sand system is necessary in order to obtain satisfactory sand.

Moisture

10. Water is added to the dry shakeout sand through a series of jets as it falls into the pug mill. The water valve is a dial type. A pressure regulator capable of holding the pressure within one pound is located in the water line ahead of the valve, the operating of this valve being done by the sand testers. The amount of water added is determined by the hourly tests for moisture, feel of the sand, past experience, and judgment of conditions in the system such as the following:

- (a) The amount of overflow sand.
- (b) The amount of shakeout sand.
- (c) The turnover of sand.
- (d) The temperature of the sand.
- (e) Extent of bridging in the storage bins.
- (f) Atmospheric conditions.

11. The effect of each of the above factors will be taken up in their respective order.

12. The amount of overflow sand is governed by the amount of sand fed to the flight conveyors in excess of that necessary to keep the molding hoppers full. The overflow sand should be kept at a minimum, otherwise it clogs both the reciprocators and screens. This material, being normally tempered sand, is somewhat wetter than the shakeout sand; consequently, the overflow should be kept proportional to the shakeout sand.

13. The amount of shakeout sand is dependent upon the rate of production. Being dry, it will require more water than overflow sand. The tester must watch the molding rate as well as the amount and kind of sand coming from the shakeouts.

14. The combination of shakeout and overflow sand represents the sand turnover per hour. Changes on the water valve are made with a change in turnover.

15. The temperature of the sand is influenced by the rate of production, the extent to which it bridges in the bins, and atmospheric conditions. As production increases, the sand temperature rises. If sand bridges in the bins, a smaller amount turns over more rapidly and becomes hotter. On hot days the sand does

not cool readily and hot sand requires more water because it dries out faster in traveling from the storage bin to the molding hoppers.

16. All of these factors must be borne in mind by the sand tester. With a good tester it is possible to hold within plus or minus 0.1 per cent of the moisture standard ninety-five per cent of the time.

17. The percentage of moisture to be carried in the sand is determined by the class of work on the system. For large work a content of 4.8 to 5.2 per cent is carried, and for small work a content of 5.2 to 5.6 per cent is carried. On hot days or with hot sand, this content may have to be increased.

Bond

18. The rebonding mixture is added with a belt feeder to the shakeout sand belt. The feeder is equipped with a variable speed reducer and a gate. With this arrangement the feeder becomes very flexible. This is advisable because all the addition sand is added in the rebonding mixture.

19. Two kinds of rebonding material can be added, bonding sand or clay. When using either of these materials, additions should be made gradually and uniformly or else balling of the clay will surely result. Raw clay should not be added, as segregation will take place.

20. A mixture of burnt sand and clay is used. This mixture is thoroughly mulled before using and is made as dilute in clay as possible without causing an undue increase of sand in the system.

21. Regardless of all precautions taken, some of the clay will ball and a suitable piece of machinery must be used to break up these balls. The revivifier tends to break up lumps of sand and clay balls.

22. Samples taken from a system using clay for rebonding after mulling for ten minutes, increased in bond from 6 to 11 lbs. compressive strength, which indicates that there was considerable ineffective clay in the sand. Mulling would eliminate this objectionable condition; however, it is impractical to install a batch-type muller in a continuous sand system where a large tonnage is used. What is needed is a continuous muller capable of handling 100 to 150 tons of sand per hour.

23. The amount of rebonding mixture fed to the system sand is regulated from the tests for compressive strength taken each

hour. For heavy work a sand strength of 4.5 to 5.0 lbs., and for light work 5.0 to 5.5 lbs., is maintained.

Permeability

24. The permeability of the sand is influenced by the following factors:

1. Fineness of the core sand.
2. Percentage of clay in molding sand.
3. Percentage of sea coal in molding sand.
4. Percentage of moisture in molding sand.
5. Percentage of dust in molding sand.
6. Temperature of sand.
7. Grain distribution.

25. Inasmuch as all of the sand additions to the system eventually come from the cores, the fineness of the core sands more or less governs the permeability of the molding sand. A high ineffective clay content reduces the permeability. Normal clay builds around the grains, producing an openness; but when the clay becomes burned to a point where its bonding qualities are gone, it powders, forming fines which clog the pores in the sand.

26. The sea coal, being fine, clogs up the sand; moreover, the ash from it will build up to a constant introducing an additional portion of fine material, and consequently, as the sea coal percentage increases, the sand becomes less permeable.

27. The moisture content of the sand influences the manner in which it rams up. Excessively wet sand will ram up tight, producing a low permeability in the mold.

28. Sea coal ash, dead clay and sand grains broken down through handling and heat, introduce dust into the sand which lowers the permeability. A large portion of this material is carried out through the ventilating stacks over the shakeouts. Fines are removed from the burnt sand by screening.

29. The temperature of the sand has a very marked influence on the permeability, cold sand being more open than hot sand. When the permeability is already rather low, it is essential that the sand be kept cool. To accomplish this, air is passed through the system sand as it falls from the vibrating screens.

30. The permeability is affected considerably by grain distribution. When the grain distribution is good, the sand is more open than a sand having a poor distribution with the same fineness number. The screening of the burnt sand permits certain portions to be accepted and others rejected. In this manner a certain grain

distribution can be built up, thereby making the sand fairly open with a relatively high fineness number.

31. Fig. 1 shows the improvement of grain distribution by the addition of screened burnt sand added through the rebonding and sea coal mixtures.

32. Sand in a system making large work has a permeability of 30 to 40, and for small work 25 to 35.

Sea Coal

33. Sea coal has many objectionable qualities, its justification being to prevent the sand from burning on the castings and to improve the finish. Sea coal and the coke produced from it weaken the sand, the coke at times picking up clay and thus forming balls which have to be screened from the sand. As previously stated, the ash from the sea coal tends to lower the permeability.

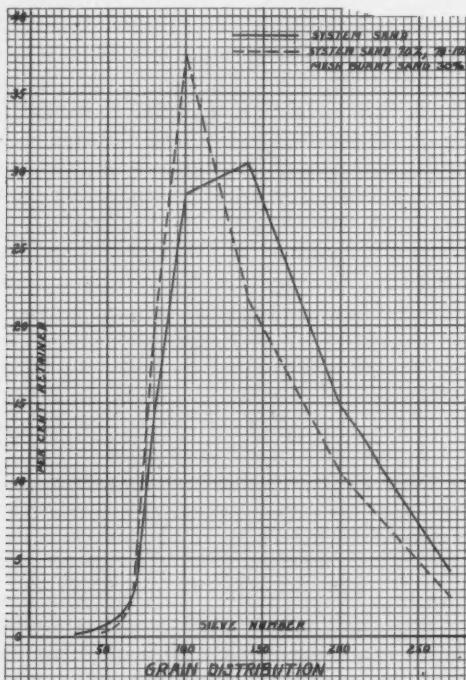


FIG. 1—CHART SHOWING IMPROVEMENT IN GRAIN DISTRIBUTION OF SYSTEM SAND BY THE ADDITION OF 30 PER CENT SCREENED BURNT SAND OF 70 TO 100 MESH SIZE.

34. To add to its objectionable qualities, sea coal is the hardest ingredient in the sand to properly control. If a quick and accurate test (similar to moisture or bond) could be made each hour, the problem would not be so difficult. Both the burning and flotation methods for determining sea coal are inaccurate and therefore unreliable, by reason of the fact that they do not show the amount of effective sea coal present.

35. The most satisfactory method of controlling the sea coal is by careful observation of castings, by accurate record of all sea coal additions to the system, and by past experience. All additions are carefully weighed, as the amount added in a day is small.

36. To obtain good results, the percentage of effective sea coal is held within narrow limits. Too little sea coal permits the sand to burn on the castings, which results in heavy sand losses from the system, hampers inspection of the hard iron castings, and causes excessive cleaning costs. In addition, the finish on the casting is not good.

37. Blows, surface checks, shrinks, misruns, cold shuts, dirty molding, and drops are common results of too much sea coal.

38. There are three ways of adding sea coal to the system, as follows:

- (a) With the rebonding mixture.
- (b) Raw.
- (c) With bank or burnt sand.

39. The first method is not satisfactory because the amount of seal coal added is dependent upon the bond strength, inasmuch as the amount of rebond added is governed by the bond strength tests.

40. The second method is satisfactory, but extreme care must be taken to prevent segregation. The sea coal must be added slowly and uniformly throughout the day, and the sand must be thoroughly mixed after additions are made.

41. The third method is most satisfactory, as it produces more uniform distribution.

42. The amount of sea coal that will have to be supplied to the sand to replace that burnt out will vary from time to time due to the following conditions:

- (a) Temperature of the iron.
- (b) Size and shape of castings.
- (c) Rate of production.

43. The higher the pouring temperature of the iron, the more sea coal will be burned out of the sand. Due recognition of this

fact must be made where the pouring temperature is changed.

44. Size and shape of the castings have a pronounced influence on the seal coal burnout. Large work produces a hotter temperature within the mold, and as a result the sand against the casting is burned back to a greater depth than with small work. The square inch area of casting exposed to the sand is not proportional to the weight, and it is possible with a rather light job to burn out considerable sea coal. It is desirable to classify jobs on certain units because of this fact.

45. The amount of sea coal fed to the system necessarily will be changed with a change in production. Accurate records of consumption show that there is a definite relationship between weight of iron poured and sea coal consumed.

46. As the volatile matter is the ingredient that produces the desired results in the mold, shipments of coal should be checked for volatile matter.

47. For large work 11 to 12 per cent sea coal is carried in the sand, and for small work 9 to 10 per cent is carried.

Volume of Sand in the System

48. For good operation, both from the standpoint of control and economy, it is desirable to maintain a constant volume of sand in the system. The various conditions affecting the volume of sand are as follows:

- (a) Effectiveness of the shakeouts.
- (b) Percentage of sea coal in sand.
- (c) Percentage of clay in sand.
- (d) Nature of the cores.
- (e) Temperature of the iron.

49. Poor shakeouts leave considerable sand on the castings, thus entailing large losses from the system. As a result, large amounts of addition sand must be made, which not only increases the costs but destroys control.

50. A change in sea coal percentage will cause a change in volume of sand. With too low a sea coal, large losses will take place; with too much sea coal a rapid sand pick-up results.

51. A high clay content causes the sand to adhere to the castings, thus producing abnormal sand losses.

52. The fact that the cores disintegrate at the shakeouts (thereby adding sand to the system) has an important bearing upon the sand volume. Change in baking conditions of the cores or the amount of core binder used in them will change the extent to which they contribute sand to the system.

53. The pouring temperature of the iron will affect the extent to which the sand sticks to the castings, thus influencing the losses from the system.

54. All additions of sand are made through the rebonding and sea coal mixtures. The variable speed arrangement on the rebond feeder allows very flexible control of the sand additions. To add more sand the mixture is made lean in clay, and to cut down sand additions the mixture is made rich in clay. For best operation the largest volume of sand possible is kept in the system, which permits it to temper and cool properly; moreover, the control is less difficult.

Factors Affecting Finish

55. A highly satisfactory casting finish can be obtained by the use of a suitable facing. When facing is used, the control of the backing sand need not be so rigid, the principal requirements being that it have sufficient bond for molding and be permeable enough to allow gas to escape from the mold. The use of facing, however, is not desirable because it slows up production and is expensive. Where facing is not used, the necessity for a high-grade sand is obvious.

56. To obtain a good finish without the use of a facing is no simple task, the nature of the sand used being such that low permeability results. The molding foreman prefers a sand that is more open than necessary to allow for discrepancies in molding practice. When a fine sand with low permeability is used, the molding technique necessary to make a sound casting becomes more exacting.

57. The finish on a casting will be no better than that produced by the coarsest particle in the sand. Most of the extraneous matter, such as clay balls, coke, rock, roots, iron shot, core particles, charcoal, etc., is removed at the vibrating screens. About 95 per cent of the sand will pass a 40-mesh screen after passing the 1/8-in. vibrating screens. Due to the shape of the sand grains, it does not matter materially whether the screen openings are 1/8 x 1/8 in., 1/8 x 3/4 in. or 1/8 x 1-1/4 in. The size of opening used is determined by the tonnage to be put through the screens.

58. The sea coal is ground to pass 99 per cent through a 70-mesh screen. Frequent tests are run to insure that this fineness is maintained, as large particles of coal cause pits in the castings. If the sea coal becomes too coarse, serious trouble may result.

59. Sea coal is used to improve the finish to a certain extent.

However, when the percentage of this ingredient in the sand becomes too great, the surface of the casting will be covered with checks and ridges, causing an undesirable finish.

60. Proper moisture content and tempering are essential to good finish. Dry sand or sand that has not tempered properly will often stick to the patterns, producing very rough castings; moreover, the sand will not mold properly. This causes rough spots at the parting line and frequently small crevices will not be filled with sand, resulting in a rough spot at that point.

61. Hot sand can be very troublesome. The hot sand coming in contact with the relatively cold pattern causes the water vapor to condense and the sand to stick to the patterns, thereby producing a rough casting. To prevent ordinary sticking of sand to the

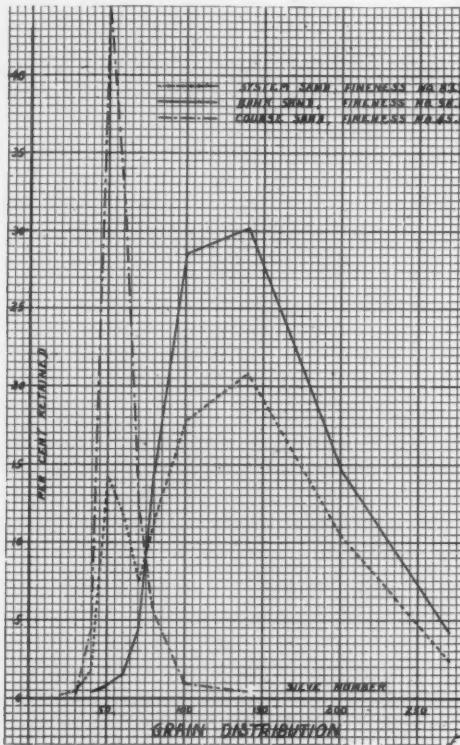


FIG. 2—CHART SHOWING DESTRUCTION OF GRAIN DISTRIBUTION OF SYSTEM SAND BY THE USE OF IMPROPER COARSE SAND FOR OPEN CORES.

pattern, a suitable spray consisting of lard oil and kerosene is used.

62. Inasmuch as the system sand has the same characteristics as the core sands, the choice of bank sands for this purpose is influenced by the desire to obtain smooth castings.

63. Ninety-nine per cent of the jobs are made without facing and have a fairly smooth finish. For a few jobs that require an exceedingly smooth finish, a facing consisting of black sand and No. 18 Helmick is used.

Relation of Cores to Molding Sand

64. It has previously been pointed out that the core sands determine the fineness of the molding sand; also, that the nature of the cores determines to some extent the volume of sand in the system. Some cores must be very open, others can be made less open. Therefore, different sands must be used for various types of cores. It is essential that the proper choice of sand for open cores be made, as otherwise the grain distribution of the molding sand will be undesirable.

65. Fig. 2 shows the effect of using too coarse a sand for open cores, with the result that the system sand has a poor grain distribution and the castings an unsatisfactory finish. Either the core mixtures must be regulated to produce a fine system sand or facing will have to be used. The use of the latter is very expensive, and considerable expenditure can be made in the selection of the proper core sands.

66. The amount of core binder used and the method of baking determine the manner in which cores disintegrate at the shake-out. It is desirable that sufficient core sand be liberated to the system to replace sand lost by adherence to the castings, otherwise the control of the molding sand becomes difficult.

67. It is apparent that control of the core sand is of equal importance with control of molding sand, but it is not the purpose of this paper to deal with core sand to any extent. Suffice it to say that the best cooperation possible between the molding department, core department and sand laboratory is essential to the production of a satisfactory molding sand.

Foundry Scrap

68. The many foundry troubles directly or indirectly traceable to faulty sand are well known to every foundryman. Broken molds, drops, misruns, cold shuts, scabs, blows, dirty molding, rough castings, etc., often are the result of improper sand. The

broken molds that can be caused by incorrect moisture content in the sand can be so high in one hour that the entire day's efficiency is destroyed.

69. At times sand is unjustly blamed for certain foundry difficulties. Improper jolting, ramming and squeezing, poor design of squeeze heads, fluctuating air pressures, gradual wear on molding machines, etc., result in troubles too often blamed on the sand. In the final analysis, the condition of the sand in the mold, not the laboratory tests, is the criterion of its fitness. It avails us nothing if the sand is perfect yet not made into the mold the same each time.

70. All complaints about the sand should be quickly investigated and the blame properly placed. The sand laboratory can be of much assistance to the molding department insofar as they can help to work out problems pertaining to the use of the sand. The laboratory should take a conciliatory attitude toward the molding department—much in the same respect as the salesman takes toward the user of his product.

Influence of Molding Sand on the Product

71. Many small jobs are coined and it is necessary that they be absolutely true to size. In the operations of straightening and machining, the castings must fit jigs perfectly, otherwise operation is slow. Faulty sand will cause pimples and fins at the parting line, rough spots, etc., which prevent the castings from processing properly. The amount of salvage grinding will be reduced with good smooth castings.

72. The sand should be of such nature that the castings will clean readily. Where castings are not milled in the hard iron and too much sand adheres to them, inspection is hampered, scrap is annealed and processed and considerable sand enters the anneal, which wastes fuel and upsets the annealing cycle. Moreover, excessive amounts of sand adhering to the castings after the shake-out cause slow cooling of the iron, and difficult annealing.

73. The finish on the casting should be the best that can be obtained, as it makes for pride in the product and insures the customer's good will.

Conclusion

74. In conclusion, it is desired to emphasize the fact that the control of sand is of major importance in the successful operation of a modern foundry.

DISCUSSION

MEMBER: How does the author arrive at the proper point for dry strength in his synthetic sand?

CHARLES MORRISON: We find that the green bond strength, if held properly, gives us satisfactory results providing the sand has good grain distribution. In other words, if we keep the sand within a certain band of grain distribution with a given green bond strength, we keep out of trouble. Therefore, we try to maintain the green bond strength as closely as possible. There has been much stress placed on the importance of the dry bond strength, but it has been our experience that we have no trouble from washes and scabs if the sand is relatively open with the proper green bond strength.

MEMBER: What is the dry bond strength of your sands? It seems to me your are using quite a strong sand.

CHARLES MORRISON: The dry bond strength runs about 40 lbs. shear. We use a relatively high green bond; that is, for the small work. For large work we run the bond considerably lower. The reason we carry it higher on the small work is because we have deep draws and have to have the strength higher in order to get clean draws.

MEMBER: In connection with the question of dry strength, has it been your experience that the dry strength shows up deteriorations faster than do the green strengths?

CHARLES MORRISON: We used to run dry bond strength and it did not tell us very much. We did a great deal of work on it, and after considerable study we decided that the green bond strength, if held properly, told us more and gave us better control and better molding conditions than the dry bond strength.

MEMBER: What type of clay is used, and how is it applied? Is it in plastic form or is it in the form of cement, and how do you apply it?

CHARLES MORRISON: We use finely ground rebonding clay. We take the burnt sand that comes from the knock-out conveyor, wet it down slightly and add the clay, and mull it for about ten minutes. It is in a semi-dry state so that it will go through the feeder readily. It is fed slowly and uniformly and is made as lean in clay as possible so as to minimize the possibility of segregation.

MEMBER: What grade of sea coal does the author use, and how does he add it to the sand?

CHARLES MORRISON: We grind our own sea coal. It is air floated, and we add it to the system raw in order to keep the number of factors of control to as few as possible. If it was not for that reason, we would add our sea coal mixed with burnt sand. We feed it into the pug mill very slowly and uniformly all day. We also weigh the amount of sea coal that goes in each day and keep the percentage within narrow limits, as we find there is a definite relationship between the pounds of sea coal added and the tons of iron poured.

G. OLSON: * Returning to the question of green and dry strength bond, I believe Mr. Morrison said it all when he stated that dry strength

* Illinois Malleable Iron Co., Chicago.

is governed to a great extent, if not entirely, by the type of sand used. In other words, when he classifies or separates his sand, he gets away from the use of fines and extremely coarse sands—his sand is very uniform in grain size. That being the case, all he needs to do is to add the proper amount of bond and he will have no trouble with scabs and cuts. That is the practical foundryman's experience.

We also know that, due to all the troubles that have been found in foundries in regard to scabs and so on, we have to investigate the reason for these defects. We have gone into the field of testing for dry strength, which of course will determine if the sand is the cause. I believe, however, that the author hit the nail on the head when he told us that all of these experiments with dry-strength bond are not absolutely necessary if one starts out with the right kind of bond.

MEMBER: The value of the sand for the job depends a good deal on the character of the clay used. One can get a clay that provides so much dry strength as to cause lumpiness of the sand, or a clay so weak as to cause cuts and washes. The proper thing to do is to get a clay that fits the job, one that does not cause lumpiness nor cuts and washes.

CHARLES MORRISON: Our mechanical setup is of such a nature that there is little likelihood of our having lumps in the sand. The shakeout sand passes through a vibrating screen and the clay is added to it at that time; very often we find that many of the lumps that are in the sand come from the clay. In other words, if the clay is screened out, it will be found that almost any clay obtainable, no matter what the brand, will have a certain amount of lumps in it. If these lumps are not added in the first place, they will never get into the sand. With our vibrating screens we remove the lumps and also break up any caked sand.

After the sand is tempered and reconditioned, it passes through the revivifier, which breaks up any lumps that are left. Thus, the sand that is delivered to the molding hoppers is very uniform and in excellent condition. We have found, in nearly every case where we have had washes or scabs, that the trouble has been caused by one of the following reasons: Mold improperly rammed, sand too wet, or permeability too low. In no case have we found that the cause was lack of dry bond strength. Where fine sand is used, it is essential that a narrow spread of grain distribution be held in order that the highest permeability may be obtained at all times.

It is my belief that too much stress has been laid upon what is used, and not enough stress on how it is used.

CHAIRMAN FRED L. WOLF: Mr. Morrison, you have quite rigid specifications on your raw coal, sea coal. Do you use any special grade? Do you notice any difference from grade to grade?

CHARLES MORRISON: We buy a coal which has a very low ash, and we grind it so that 99 per cent passes the 70-mesh screen and 90 per cent of it passes the 200-mesh screen. We run tests frequently to be sure this screen fineness is maintained.

CHAIRMAN FRED L. WOLF: Is that a high-volatile or low-volatile coal?

* Ohio Brass Co., Mansfield, Ohio.

CHARLES MORRISON: About 33 per cent volatile.

MEMBER: Mr. Morrison, I would like to know if you have associated a time temperature with a particular kind of core binder? You mentioned some optimum relation of time and temperature in making the cores. While I recognize that it varies with the different types of work, it would be interesting to know if you have any figures to give us. You spoke of baking the cores longer and having a lower binder content. What kind of binder have you found most valuable?

CHARLES MORRISON: In our particular case we use core oil principally. We find that we can get along with very little oil if we want to bake it a long time. Also, getting back to grain distribution, the sand that has the best grain distribution naturally requires less oil because it has less surface area to cover, and the coarse sands use less oil because they have less surface area to cover.

Of course, it is not practical to classify core sand, for the simple reason that it is too costly; but what one can do is to blend two bank sands in such a way as to obtain an intermediate product of the desired characteristics. Of course, the amount of baking time and the amount of oil used is dependent on the production rate. In other words, if there are two core ovens and the operator has to get just so much out of them, one might have to vary the practice in order to produce the necessary quantity of work, whereas, if a longer time were available, one might do it another way.

MEMBER: We have found that for every inch in thickness of the core, one hour produces the same core with a ratio of linseed oil of about 1 to 100. I thought perhaps you had something of that kind in mind.

CHARLES MORRISON: We are doing a lot of work in core baking; but, as I mentioned in this paper, if we were to put it in the paper it would have been so long as to be tedious. It really calls for a separate paper. However, there are so many factors in core baking that if they were to be taken into consideration we would have to determine the relationship in that manner. It is not at all definite because the practice varies from day to day.

The practice varies with the time. We do try to keep the relationship of the manner in which the cores are made fairly stable. We do not want one core to come along and break up and give up sand into the system and the next core to give no sand into the system, because if that happens we have lost our control.

MEMBER: Mr. Morrison, you say you are using two bank sands and that you blend them yourself; you have found that that is better in your case than buying the sand already blended from the producer. I ask this question because it conflicts with what was said a while ago, that we like to have one grain size. What do you mean by saying you use two bank sands and mix them?

CHARLES MORRISON: We all know some cores have to be open. If they are made tight, the castings will be scrap. Consequently, there is no way of getting out of making some cores open. In order to make them open, a bank sand must be used which is relatively open. Naturally, that bank sand will probably be somewhat coarser than the bank sand that can be used on cores that do not have to be open.

In order to get a sand that falls within a reasonable grain distribution range, we pick two bank sands which, when blended together, will form a material that has a lot of grains available between our sieves so that when we screen it we get the maximum recovery between those sieves.

MEMBER: Do you screen the sand before you use it?

CHARLES MORRISON: We do not screen our bank sand before we use it; we only screen the burnt sand. To screen the bank sand would be rather a costly proposition. In choosing these bank sands, we choose them somewhat on the basis of their grain distribution; we try to obtain sand that falls as much on one screen as possible. We carry the fineness number of that sand as high as possible and still maintain the given openness of a certain core that has to be open. On the other hand, we go to the fine side and try to bring a fine sand on a very narrow band. Thus, when we blend the two together and dump the blend into our system sand, we get a sand that has the proper grain distribution. In other words, our hands are somewhat tied because, in spite of the fact that we might want to make all our cores out of the same sand, we cannot do it. We have to make some out of a coarser grain in order to have it open enough so that the casting will not blow or hot crack.

MEMBER: The author says that mulling will eliminate this objectionable condition of lumps in the clay bond. However, that may be impracticable. You may not be able to install a muller capable of handling 100 to 150 tons in this short time. Has Mr. Morrison had any experience with a continuous mulling system? Several such mulers can be put into the flow sand system with the sand being mulled as it comes through the tunnel on a moving platform. It can be mulled at any given rate.

* CHARLES MORRISON: We have gone into the mulling of our sand quite extensively and, after considering all the factors involved—cost of power, results, etc.—we have come to the conclusion that a batch-type muller would not be satisfactory. However, a continuous muller that would not require too great an outlay of money, if it is economical to operate, might be practicable.

The figures in this paper were quoted some time ago. Our sand at the present time does not carry anywhere near the bond it used to. We have found ways and means of dropping off the clay content, although one never can get down as low on the clay without mulling as with mulling.

MEMBER: There seems to be considerable interest in this question of dry bond. I wonder if the speaker has found from his experience whether there is any difference between a plant having a continuous system in which there is no storage of molds and in which they would be drier, and a plant having no such system and in which there would be considerable range of moisture.

CHARLES MORRISON: I believe it would have a more pronounced influence on the sand. In other words, the more discrepancies there are in the sand, the greater the influence of any single discrepancy. Where the tendency is for the sand to become wetter or drier or for the strength to vary from one to the other, or where the molds set for a considerable length of time, very likely the foundryman will be concerned to a great extent with the dry bond strength.

We are not bothered with these factors; that is, the molds are poured off quite rapidly and, as near as we are able to determine, if we maintain the green bond and keep the sands in proper condition, there seems to be no advantage in running the dry bond strength. For instance, we have certain clays that claim to have a higher dry bond strength than those of another class. They have, but from a practical standpoint we get no better results with them than we do with a clay that has a lower dry bond.

Handling and Conditioning of Sand in a Small Gray Iron Foundry

By MAX A. AMOS* AND ROBERT W. GRACE,† MUSKEGON, MICH.

1. In the Fall of 1930 the Standard Automotive Parts Co., Muskegon, Mich., was experiencing considerable trouble with molding sand. After a thorough study of the situation it was decided that the sand must be conditioned. Upon viewing the situation from the standpoint of using mechanical equipment, all types were studied to fully determine which type would be best suited to efficiently and economically handle production.

2. The production factors involved were, first, sand requirements at peak production; second, sand requirements at approximately half production; third, any other condition by which part of the equipment would be required and which still would not make it necessary to operate the entire system, as would be necessary if a continuously operating handling system were used, since there would be no economy in having a sand-handling system not flexible enough for partial operation on a production schedule below normal. Continuously moving machinery in a foundry wears, even if it is not working to capacity.

3. In addition, business conditions did not warrant an expenditure of capital for new buildings, so that a system had to be selected which would adapt itself to the existing building.

4. Before explaining the system installed, it is proposed to discuss the type of castings made in this plant, and how the sand was previously handled.

Types of Work Produced and Original Conditions

5. The product of this foundry is cast-iron valve guides. These vary in weight from one ounce to over one pound. They are not cast solid, as is usually the case, but are cored. Many of

*Metallurgist, Standard Automotive Parts Co.

†Foundry Consultant.

NOTE: This paper was presented and discussed before the joint A.F.A.-A.S.M.E. materials handling session at the 1933 Convention of American Foundrymen's Association.

the cores are about the size of a lead pencil and are made from lake sand and linseed oil, and reinforced with a wire. On all surfaces to be machined, 1/32 inch of stock is provided for finishing, an operation performed in the plant's own machining department.

6. The sand was formerly prepared at night by men who shook out the molds and moved the castings and wet down the shakeout sand with a hose. Then, by means of a small sand mixer, the sand was thrown in a heap as close to the molding machine as possible, ready for the molder to start work on the following morning.

7. These men had no supervision, and it was found on several occasions that only a small portion of the sand actually passed through the mixer. As a result, the sand ranged in moisture content anywhere from 6 per cent to 10 per cent, and many times the following day some molder would lose a number of copies and a lot of his patience before realizing that he had to cut over his heap in order to get the sand in a workable condition.

8. This naturally caused low production and the molder had a great deal of unnecessary work to do that the night men had been paid for and for which the molder would get no pay. This condition also reduced his opportunity to make premiums on a piece-work rate, and his scrap for the day increased greatly due to the poorly conditioned sand he had used in his molds before cutting over the heap.

9. A large molding machine also is used for making multiple stack molds on large production orders. On this machine also considerable trouble was encountered due to the condition of the sand. The sand must be very uniform, a definite volume is required for each operation, and it must squeeze to the same depth every time. Sand of varying temper fails to do this.

10. Any foreign material in the sand also causes trouble because, if it comes in contact with the automatic sprue cutter, it generally breaks the cutter and it then becomes necessary to replace the broken part, which naturally stops production.

11. The general condition of the sand became such that a considerable amount of scrap was found, due to scabs, blows and sand inclusions. These defectives were very expensive, because most of them were not discovered until well along in the process of machining. It is well known that scrap which costs 40 cents per pound is very expensive melting stock.

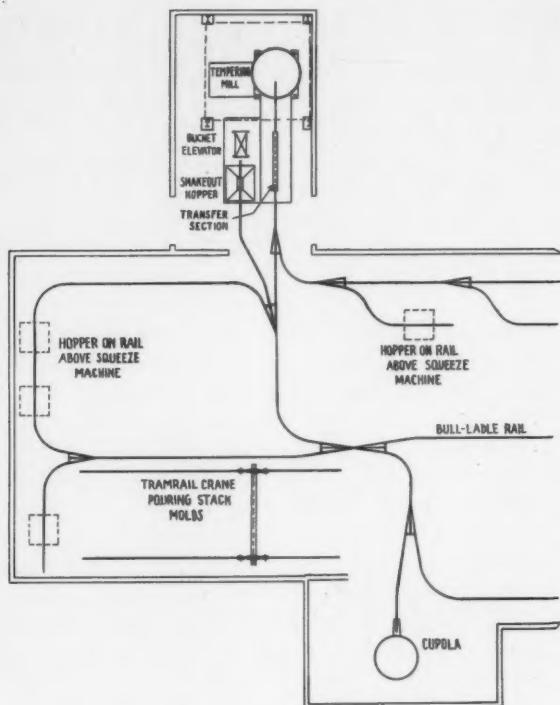


FIG. 1—LAYOUT OF SAND CONDITIONING EQUIPMENT AND TRAMRAIL SYSTEM.

Requirements Considered

12. Before any decision was reached as to how conditions could be bettered, a number of methods were considered. The first decision was to remove foreign metallic material from the sand; this was considered best accomplished with a magnetic separator. The next step was the tempering of the sand to put it in better condition for the molder's use.

13. A number of different types of tempering mills were considered, but finally it was decided that a sand mill of the muller type was best adapted to the condition. With this, additions of new sand and water could be accurately controlled. This settled the question of conditioning.

14. The problem of handling sand was next considered, and it was decided that the most practical way would be to use an over-

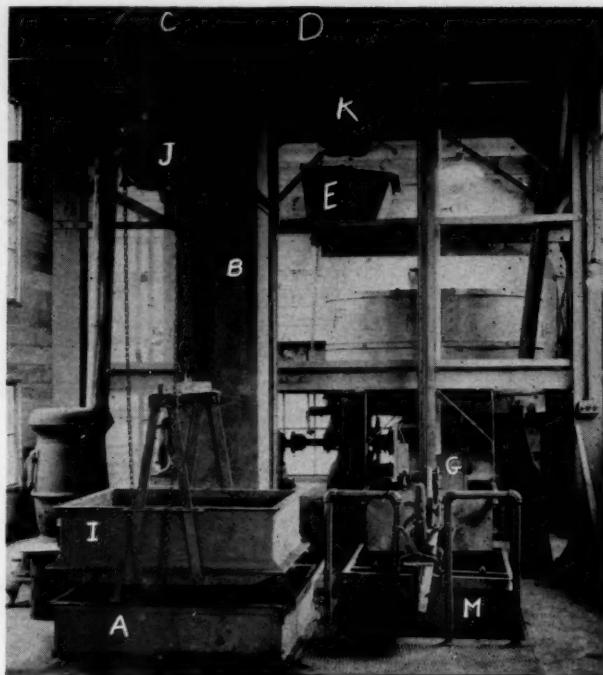


FIG. 2—VIEW OF SAND CONDITIONING ROOM AND EQUIPMENT.

head tramrail system with sand hoppers suspended on the load bars of hand-propelled carriers, which could be moved from the conditioning unit to the molding machines. In addition, it was decided to study the possibility of reducing the cost of handling metal to the pouring floors, and the cost of handling castings to the shakeout and cleaning department. This was accomplished very satisfactorily and at minimum expense, as it was found that the tramrail equipment for handling the sand could be extended at a nominal cost to take care of the pouring and shakeout operations.

General Layout

15. Fig. 1 shows a plan layout of the foundry. The sand conditioning equipment (Fig. 2) consists of a shakeout hopper *A* with vibrating screen, a magnetic separator unit, a bucket el-



FIG. 8—MOLDING MACHINE STATIONS WITH SAND CONVEYOR HOPPERS OVER EACH MACHINE.

evator *B*, a suction fan *C* for removal of fines, a 90-ton sand-storage bin *D*, a measuring hopper *E* and a muller-type mixing mill *F* with "squirrel-cage" aerator *G* attached.

16. On the East side of the foundry is a wing (see Fig. 1) in which all sand-conditioning equipment is located. It is centralized for ease of operation and economy, an addition being necessary to house this department. On the North end and East side of the foundry the squeeze machines are located, with the movable sand hoppers above. This arrangement is shown in Fig. 3. In the Northwest corner is located the large molding machine which makes the multiple stacked molds on large orders. The floor behind this machine is served by a tramrail crane for pouring and shakeout.

Operating Procedure

17. As to the operating procedure, the sand first enters the conditioning equipment at the shakeout hopper *A*. It is brought here by what is called a shakeout bucket *I*, which is suspended from the tramrail by an extended hand-wheel chain hoist *J*. To deposit the load, the hoist is lowered, and as the sides of the shakeout hopper and the bucket come together, the bottom of the bucket falls away, being hinged at the sides, and allows the sand to flow through. This is shown in Fig. 2, which also gives a good view of the conditioning equipment.

18. In the shakeout hopper *A* is a vibrating screen where

the large pieces of iron such as sprue, parts of a gate or pieces of spilled iron are taken out. If this material were allowed to pass through, it would damage the belt on the magnetic separator unit.

19. Next, the sand goes over the magnetic separator, where core wires, shot iron and all magnetic materials that have accumulated in the sand while in the foundry are removed.

20. Next the sand, now thoroughly cleaned, falls into the boot of an enclosed bucket elevator *B* and is raised to the 90-ton steel storage bin *D*, where it cools and is ready for conditioning the following morning. The bin capacity is greater than the amount of sand used in a day's run at peak production. It is fitted with a hopper bottom and double-swing gate *K* so that sand may be easily withdrawn.

21. Directly below and attached to the bottom of the storage bin is a measuring hopper *E* and below this is the tempering mill *F* with aerator *G* attached at the discharge gate. From an operating floor about five feet above the general floor level, the operator controls the gate on the storage bin, the measuring hopper, the mill and the aerator.

22. In preparing a batch of sand, the measuring hopper is filled with a thousand pounds of sand and the sand is then dumped into the mill, where water and new sand are added to keep the sand in proper physical condition. The water is sprayed in from a ring of jets located at the top of the mill's shell and is measured by a gasoline-type meter which registers gallons and fractions of a gallon.

23. Hourly, a sample of prepared molding sand is taken to the laboratory, where it is tested for moisture, permeability and bond. Occasionally a screen test is run on the sand. If a variation from standard is found, the operator is told what changes to make to obtain the desired results.

24. The batch is mulled about three minutes and discharged through the aerator, the freshly prepared sand falling into the movable hopper *M*, which is suspended on the hand-propelled tramrail carrier. This hopper has a capacity of 2000 lbs., or two mill batches. It is fitted with a clam-shell gate. The tramrail is so arranged that the hopper can be placed directly under the aerator to receive the sand. After the second batch has been run into the hopper, it is moved onto the tramrail two elevation transfer section and electrically hoisted to the upper tramrail level. This is shown in Fig. 4.

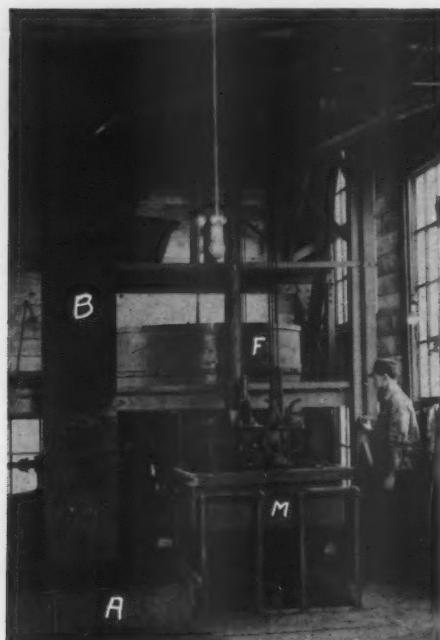


FIG. 4—ELEVATING CONVEYOR HOPPER *M* TO TRAMRAIL DISTRIBUTING SYSTEM.

25. From there it is easily pushed by hand along the molding-sand track and left momentarily at a point beyond a switch which connects with a storage-line track above the squeeze machine requiring sand. The empty hopper is then removed and the loaded hopper pushed into position on the track above the particular squeeze machine. This transfer of hoppers requires very little time, and the operator then returns to the mill with the empty hopper for another load. Thus, a ton of sand has been conditioned and transported.

26. This places the sand over the molding machine so that it may be dropped into the flasks as needed. Shoveling has been eliminated, except between loads. This has worked out very satisfactorily because it is not possible to avoid spilling some sand on the floor when making a mold, and what is spilled can be used to make a mold or two while the empty hopper is being replaced by a full one. In this way the molder is not held up for lack of sand.

27. Seven molders are employed on 12 x 17 flask work, each making 180 molds per day; one molder on the molding machine with steel flasks for stack work, who makes 350 molds per day; and one job molder, making 50 miscellaneous molds per day. Complete molds are set in rows; after the molder has put up his floor, he pours it off and is through for the day. Under present conditions, the output of castings is less than five tons daily, requiring the handling of 42.7 short tons of sand.

28. The molds are allowed to steam for about two hours to dry out the sand before shaking out. This aids in getting the sand through the conditioning equipment, and the fan does a better job of removing fines. Also, the sand is cool when used the following day, eliminating sand sticking to cold patterns.

29. The night men come in at the proper time and begin shaking out the molds. The shakeout bucket is brought to the floor to be worked on, and the molds, which are on bottom boards, are picked up one at a time and dumped into the bucket. The castings are hooked out and dropped into another container. When the shakeout bucket is filled, the load is pushed to the shakeout hopper and dropped. This completes the cycle of operation.

Removing Fines from the Sand

30. The first and only great problem encountered in operating this system was that of fines. In milling the sand that had been in heaps previous to the installation, it was found that the permeability dropped sharply and the bond increased. Not stopping to consider that milling the sand broke up compound grains which had formed in the heaps due to excess water, sand with high permeability and low bond was added in an effort to remedy the situation. This reduced the bond but affected the permeability very little.

31. On making screen analyses of the sand it was found that the sand had a large percentage of fines, which were increasing daily, no doubt due to milling. This not only gave a low permeability, but the sand was brittle when it was made into molds.

32. The conclusion then was reached that there was no way out but to remove the fines. This was accomplished by means of a suction fan with large volume and low velocity. The inlet was placed in the casing of the bucket elevator, midway between top and bottom, to pick up the fines where they first floated in the

air. This proved successful and the sand soon returned to a normal condition.

33. Controlling the consistency of the sand proved to be a great help, as scrap now is only a fraction of what it was formerly. The moisture content of the sand seldom varies more than two-tenths of one per cent, plus or minus. The compression bond readings of the sand are almost constant and vary only as the moisture content varies, and seldom do the molders complain that the sand is at fault. The castings seldom show sand inclusions, and then the trouble usually is due to carelessness. There are very few copies dropped, and troubles on the large molding machine are greatly reduced.

Cost Comparison

34. Following is a comparison of the cost of handling 42.7 short tons of sand with the system. The allowance for depreciation, interest on investment, insurance, taxes, light and power totals \$1744.44 per year, or \$5.81 per day when figured on the basis of 300 working days. The direct labor cost of the man handling the mill and shifting the hoppers is \$3.15, and the labor of the shakeout men is \$11.07 per night. As mentioned before, the shakeout men get the sand back to the receiving hopper of the storage bin and leave the castings knocked off and the sprues in separate containers, so that the molders are ready to start work in the morning immediately upon their arrival. The total cost of the above shakeout labor and mill labor, plus the overhead, amounts to \$20.03 per working day, or \$0.468 per ton, or \$0.012 per mold.

35. Before this equipment was installed, the shakeout men cut the sand with a sand cutter, tempered it with a hose and left it in piles for the molders, and were paid \$24.90 per night for this work. Dividing this amount of 42.7 tons of sand handled gave a cost for preparing the sand, under the old method, of \$0.583 per ton, or \$0.015 per mold.

36. Thus, milled and carefully prepared sand now is obtained in a very fine condition for about 12c per ton less than for the very poorly prepared and unsatisfactory sand formerly obtained under the old method. Any foundry executive will appreciate that, in addition to effecting a saving of 12c per ton by properly conditioning and handling the sand, the saving made in reduction of casting losses, etc., is an important factor. Also, about three carloads of sand are being saved per year, since only

about 10 per cent as much fresh sand is being used now as was formerly used under the old arrangement.

37. The system as now installed is capable of taking care of three additional molders, or an increase of 33 per cent in foundry activity, which would run the sand requirements to 57.27 net tons per day. To do this, the plant would expect to add one more man who would transfer the loaded and empty hoppers, leaving the present man devoting all of his time to the loading and operation of the mixer. This would increase the cost of handling and preparing sand to 47c per ton, or \$0.01226 per mold, a very slight increase over the cost with the present layout.

38. However, this additional man would be in a position to deliver sand to more molders, so that the total cost per ton would drop if it was found necessary to employ over twelve molders. It probably would be found necessary to install a larger mill if more than twelve molders were employed, and if this were done, the present sand-storage hopper with the new mill would have enough capacity for sixteen molders, so that the total cost of handling and preparing sand would be brought down below 40c per net ton.

39. The authors desire to say, in conclusion, that since operating this system we have been very much pleased with its performance. General conditions of the foundry have been materially improved from the standpoint of cleanliness and orderly arrangement. Production per man has increased and there seems to be an element of satisfaction among the molders that did not exist before. It is our opinion that a system such as this could be used in a great many foundries, regardless of the work, as it is very flexible and could be installed in units which would lend themselves to varying production schedules.

DISCUSSION

M. A. AMOS (*In presenting his paper, Mr. Amos added the following remarks*): Not many of these castings weigh one ounce; only a very few do. The average casting weighs between 3 and 4 ounces.

We have very large additions of core sand through the cores. Every casting that is made has a core through it, and although the cores are not very large, neither are the castings. Thus, we get considerable adulteration from core sand, and we find that the muller bonds this core sand that goes into the shakeout sand much better.

I would like to amend the statement about flask sizes. Although most of our flasks are 12 x 17 inch flasks, we also use 12 x 13 inch, 12 x 18 inch and 10 x 18 inch flasks; the majority of the work, however, is made in 12 x 17 flasks.

We have been using about 10 per cent as much sand as we did under the old conditions, but we have only had about 40 per cent as much business. Figuring our saving, we probably are using about 25 per cent as much sand as we did under the old arrangement.

If we had had the larger mill in the first place, very likely one man could take care of the 12 molders alone, because it would only necessitate one batch of sand per hopper, where now there are two.

About the saving of scrap losses, I looked back through the records for a period of several years and found that previous to the installation of this system, our average scrap as found in the machine shop was 7.4 per cent; for the period since it was installed, about 2 years, the average has been 3.5 per cent. We hope to get this still lower.

MEMBER: Why do you leave the castings in the sand for such a long time? Is it a matter of the labor available, or of improving the castings?

M. A. AMOS: The principal reason was to dry the sand to get the fines out, although, after we had the fines under control, we did not follow that practice through. In fact, at the present time, when the weather is hot, the molds are sprayed down before they are shaken out, so as to make it cooler for the shakeout men. In the first place, when the fines were bothering us considerably, the molds were allowed to dry out thoroughly, and were pretty dry when shaken out.

MEMBER: How do you handle the molds when they leave the molding machine?

M. A. AMOS: They are poured in floors; they are walked out.

MEMBER: Would not that system pay for itself still more if you had something on which to lay the molds?

M. A. AMOS: Possibly, if we could pour on a continuous basis; remember, however, that we are only pouring five tons and melting it with a cupola.

MEMBER: I should think, from my observations, that a man spends an average of half a minute per mold running them out, and walks about 10 or 15 miles a day. We have found that is quite a fatigue element.

M. A. AMOS: Some day we may possibly have a continuous foundry, which I think would work much better with this system.

MEMBER: The reason why I asked that question is that in our own foundry we pour twice a day and we use gravity rolls. We allow a man to stack up on the floor, and we figure that saves him about 25 per cent of his time.

M. A. AMOS: How many tons do you pour daily?

MEMBER: Up until a month or two ago, I believe we were pouring about five tons; ours is a malleable plant.

CHAIRMAN J. B. WEBB: Mr. Amos, what do you call a "fine"—what mesh screen?

M. A. AMOS: Material that will pass a 270 mesh. We do not eliminate this material entirely; probably, the way our sand runs, we have about 7 per cent of fines. We carry that much, not because we need to, but because we have had no difficulty with that much.

CHAIRMAN J. B. WEBB: What percentage of new sand, roughly, do you have to add to the old batch? Of course, that would vary with what you mix.

M. A. AMOS: That is a hard question to answer; it varies. One special feature in connection with this installation is that we have different types of jobs which require different sands, and this is made up to order as each individual molder wants it. If we have a job that requires a particularly strong sand, he gets it.

CHAIRMAN J. B. WEBB: What do you mean by strong sand?

M. A. AMOS: High bond, high strength—that is, more clay. Also, once in a while the job molder makes a large casting for which he needs a more open sand, a sand of higher permeability, and that is easily obtained.

A. F. ANJESKEY: Mr. Amos, what do you mean by "fines?" You mentioned in your paper that the fines were probably caused by the mulling operation. Didn't you have trouble with fines before you actually went to mulling?

M. A. AMOS: No, we did not; then, we had an entirely different condition. There is not much mixing action to most small sand mixers or sand cutters, and a good deal of the time the sand was worked very wet in order to get it into condition so the molders could use it. Also, there was a compounding of the grains, that is, several grains cementing themselves together, and much of the burnt clay no doubt entered there. Upon mulling, these compound grains were broken up into individual grains; then, of course, this burnt clay loosened, and there were broken-off pieces of small silica sand which would be below the 270 mesh. These were the fines which we removed.

MEMBER: Is there a large percentage of fines? Have you any idea how much weight of fines per day you get?

M. A. AMOS: We reduced the amount from about 21 per cent down to 7 per cent. Those figures are by weight.

* Jervis B. Webb Co., Detroit.

† Cleveland Crane & Engineering Co., Tramrail Div., Cleveland.

MEMBER: Do you use a slip flask or a steel flask, and do you have slip jackets on pouring?

M. A. AMOS: We use a cherry snap flask. On some of the heavier jobs, where the pattern comes close to the edge, we use slip jackets, although the majority of the jobs are poured without jackets. The larger tappet guides and similar jobs are poured with jackets. I might add that there are many jobs on which we formerly used these jackets, on which we do not use them now since we conditioned the sand.

MEMBER: You mean that you put them in a larger flask?

M. A. AMOS: No, we do not use larger flasks, but now the sand holds up better and the metal does not break out. Formerly, there were many jobs on which it was thought necessary to use the slip jackets in order to pour; on these the jackets now are being entirely left off. The bond strength of the sand does run considerably higher than it did before.

MEMBER: In connection with the fatigue of the molder in walking from the location of the hopper back to the end of the floor, do you think it would be possible, if the hopper was run parallel to the floor (that is, run the length of the floor), that that would relieve the fatigue to some extent?

M. A. AMOS: We have considered having the molding machine run on the tramrail track and the hopper following it up, although we have not done anything along those lines as yet. Possibly, when business conditions improve, that will be followed up. I believe it would work out very nicely.

MEMBER: When it comes to pouring off, you must have to employ extra help, don't you?

M. A. AMOS: The molders pour off their own floors, but we call in shifters from the shop. Of course, we have plenty of help in the cleaning room that goes to the foundry during the pour-off. The castings all have to be snagged on both ends, and there is considerable snagging labor.

MEMBER: Is there any trouble in connection with the sand arching in the hoppers over the molding machines?

M. A. AMOS: Yes, we have some difficulty there. We have a very poorly designed hopper for the flowability of the sand; due to the low headroom in the foundry, we have a fairly steep angle on the hoppers. To some extent we cured this trouble by placing chains in the hoppers, the chains being suspended from the top with springs. This seems to break down the sand, and it flows out rather well; before this was done, however, we had considerable difficulty.

The chains are quite similar to the side chains of automobile skid chains. They are suspended by springs from the cross-bars across the top of the hopper and are fastened at the bottom of the gate, which is of the clam-shell type. When the gate opens up, the chain saws through the sand. Also, there is considerable shock when the gate is closed, as there is considerable tension in the chain springs, and closing the gate jars the whole hopper.

MEMBER: Is your main storage bin outside, or in the roof?

M. A. AMOS: We have a separate building unit in which all this conditioning equipment is located. There is a building built around this, and the roof to the tank is the roof to the building. We have no trouble from freezing. The sand is dry when it goes in, and there is plenty of heat inside. In fact, during dull periods when we had occasion to run one day a week, and during very cold weather when the sand stood from Monday night until the following Monday morning, there was no bridging due to freezing.

MEMBER: Are you using a natural bonded sand or synthetic sand?

M. A. AMOS: We use an entirely natural bonded sand at present. We had a stock of sand on hand which we have not exhausted. I expect that when that is gone we will probably go to a synthetic sand.

Effect of Silt on the Bonding Strength of Sand*

By H. RIES† AND R. C. HILLS, ** ITHACA, N. Y.

1. As is more or less well-known, the "clay" separated in the A.F.A. fineness test is a mixture of true clay particles and silt. In the elutriation method these two particle sizes are supposed to be separated, although the accuracy of this separation may sometimes be questioned, since the factor of water temperature usually appears to be neglected. The question, however, arises whether clay and silt should be determined collectively in a fineness test and regarded as bond.

2. Should it be desirable, for practical or other reasons, to determine true clay and silt separately, the writers do not advocate using the elutriation method, as there are others which are accurate, shorter and require much less water. Moreover, they permit the use of distilled water, which is essential for accurate work.

3. There is no doubt that true clay serves as a bonding medium; but can we be sure that silt does? If the silt has no bonding value, or much less strength than the clay particles, it might explain why two sands with approximately the same A.F.A. clay content and about the same grain distribution, may differ in their bonding strength.

4. In order to obtain some information on the effect of silt, the writers prepared a number of synthetic mixtures consisting of clean silica sand, clay and silt. The character of the ingredients used is given below.

5. *Clay.* The material selected for this series of tests was a bentonite containing 99 plus per cent colloidal matter as determined by both the pipette and hydrometer methods.

*This investigation was carried on with the aid of a Heckscher grant from Cornell University, Ithaca, N. Y.

†Department of Geology, Cornell University.

**Department of Geology, Cornell University.

NOTE: This paper was presented and discussed before one of the sand sessions at the 1933 Convention of American Foundrymen's Association.

6. *Silica Sand.* Two different sizes of Ottawa silica sand were used. For the one series a 20-30 mesh (99.5 per cent on No. 30 sieve), round-grained sand was selected. For the second series, Ottawa banding sand was used, the sieve test of which gave the following:

On 40.....	0.09 per cent
On 50.....	0.28 per cent
On 70.....	1.85 per cent
On 100.....	37.66 per cent
On 140.....	28.56 per cent
On 200.....	18.90 per cent
On 270.....	7.98 per cent
On Pan.....	4.72 per cent
	100.04 per cent

7. *Silt.* The grain sizes usually included under this term range from 0.01 mm. to 0.05 mm. in diameter, although it is probable that the smaller size is a little too large and that 0.008 mm. grains should be included.

8. Since it was impossible to obtain any natural product whose grains were all included in this size range, the next best thing was to select materials which had a preponderance of the proper sized grains, and to separate as completely as possible all those which were larger or smaller than the desired size. Two samples of silt were therefore prepared artificially.

9. One of these samples was obtained from a grade of ground silica sand known as Jasper silica, kindly supplied to us by the Pennsylvania Sand Co. This was composed of clean silica grains. The second type was separated from the VanHoesen sand supplied us by Whitehead Bros. The grains of this silt have a surface film of iron oxide, and the two types were used in order to determine if possible whether the iron-stained grains showed greater adhesiveness than the clean silica grains.

10. Each of these materials was stirred with water to which some NaOH had been added, and then allowed to stand until everything but clay had settled out, the process being repeated a number of times. After removal of the clay, the sand grains larger than silt were separated.

11. Detailed measurement of the particle sizes of these two silts, as determined by the hydrometer method, are given in Table 1. These analyses indicate that the grains of these two silts are practically within the proper range for that size material.

Table 1
PARTICLE SIZES OF SILT SAMPLES TESTED

Silica Silt		Van Hoesen Silt	
Percent.	Size, mm.	Percent.	Size, mm.
100.0	finer than 0.053	100.0	finer than 0.053
78.1	" 0.050	96.0	" 0.050
70.0	" 0.045	91.0	" 0.043
65.0	" 0.042	68.8	" 0.035
51.1	" 0.038	51.2	" 0.029
45.9	" 0.028	40.0	" 0.025
40.5	" 0.024	32.2	" 0.022
35.1	" 0.022	25.6	" 0.021
32.4	" 0.018	20.8	" 0.019
25.6	" 0.015	16.0	" 0.018
16.1	" 0.012	14.1	" 0.015
11.3	" 0.010	12.8	" 0.012
0.2	" 0.008	10.1	" 0.010
		0.1	" 0.008

Method of Mixing.

12. The ingredients were first hand-mixed dry. Following this they were mixed dry for 5 minutes in a laboratory muller, after which the moisture was added and the mixing continued for another five minutes. The samples were then placed in closed jars and allowed to stand for 24 hours.

Testing.

13. When ready to test, a sample was taken for moisture determination, and other lots were used for compression, tensile and permeability tests. The compression tests were run on both the Adams and Dietert (motor-driven) machines, and the figures given for both compression and tension are in all cases the average of 3 determinations. Permeability was determined by the standard A.F.A. method. The tests on the two compression machines were in very close agreement, although the strength determined by the Dietert apparatus was invariably slightly higher. All the mixtures were tested at 3 moisture contents.

Mixtures Used.

14. The mixtures used contained the following:

10% silt	1% bentonite clay
9% silt	3% bentonite clay
7% silt	5% bentonite clay
5% silt	5% bentonite clay
	10% bentonite clay

This series was used with both sands and both kinds of silt.

Results.

15. The results of the tests are shown in the accompanying graphs, Figs. 1 to 8, inclusive, on which are plotted the compression, tensile strength and permeability of the mixtures tested at different moisture contents.

16. Comparing Figs. 1 and 2, representing mixtures contain-

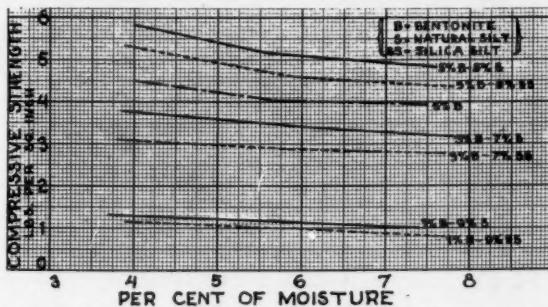


FIG. 1—STANDARD SAND. COMPRESSIVE STRENGTH AS AFFECTED BY SILT AND MOISTURE CONTENT.

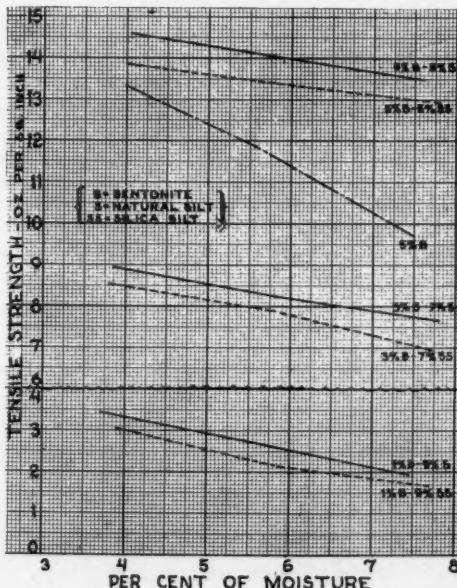


FIG. 2—STANDARD SAND. TENSILE STRENGTH AS AFFECTED BY SILT AND MOISTURE CONTENT.

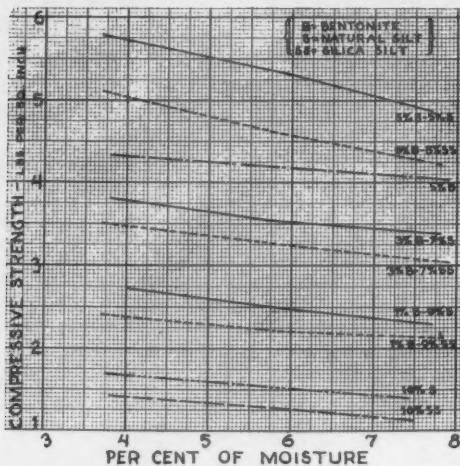


FIG. 3—BANDING SAND. COMPRESSIVE STRENGTH AS AFFECTED BY SILT AND MOISTURE CONTENT.

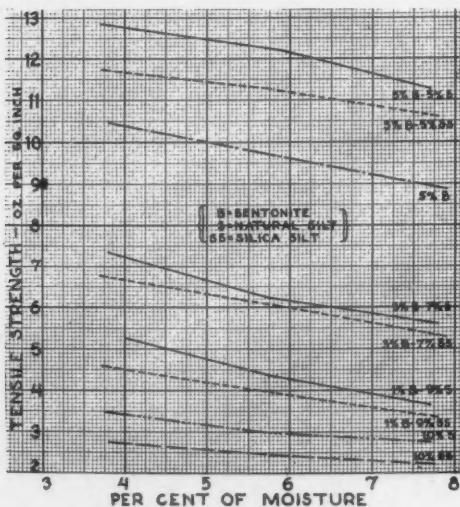


FIG. 4—BANDING SAND. TENSILE STRENGTH AS AFFECTED BY SILT AND MOISTURE CONTENT.

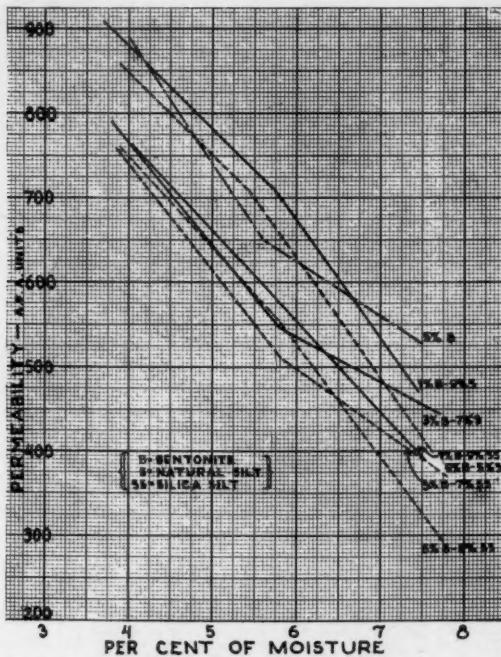


FIG. 5.—STANDARD SAND. PERMEABILITY AS AFFECTED BY MOISTURE CONTENT.

ing 20-30 sand, it will be noticed that both compression and tensile strength fell off with an increase of silt, but that the 5-5 clay-silt mixture showed a higher strength than the sand with 5 per cent of clay alone. The 10 per cent silt mixture had no green strength. In every case, the natural silt grains whose surface was coated by a film of iron oxide gave higher strength than the clean silica grains.

17. Comparing Figs. 3 and 4 representing the banding sand mixtures, we find the results not quite as expected. The compression tests with the banding sand were similar to those already obtained with the coarser sand, except in the 1 per cent clay-9 per cent silt mixtures, which ran at least a pound per square inch higher. The 10 per cent silt with banding sand gave about $1\frac{1}{2}$ lbs. compression strength, as against zero for the coarser sand; the mixture also had sufficient tensile strength to be tested.

18. Referring to the permeability tests as shown in Figs. 5

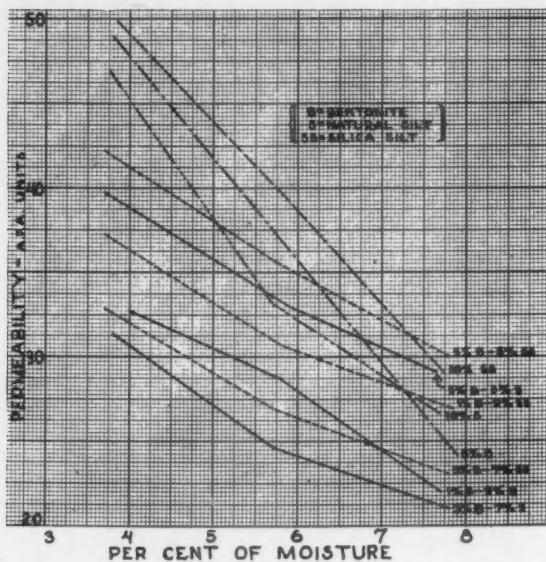


FIG. 6—BANDING SAND. PERMEABILITY AS AFFECTED BY MOISTURE CONTENT.

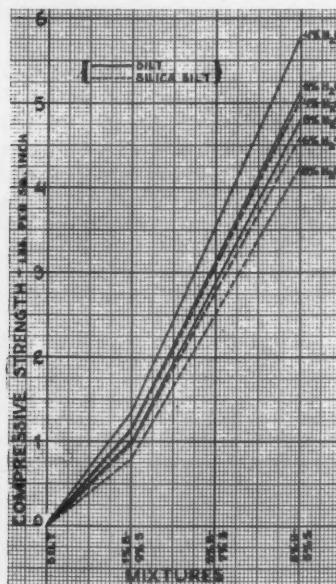


FIG. 7—MIXTURES OF STANDARD SAND, SILT AND CLAY. COMPRESSIVE STRENGTH AS AFFECTED BY MOISTURE CONTENT.

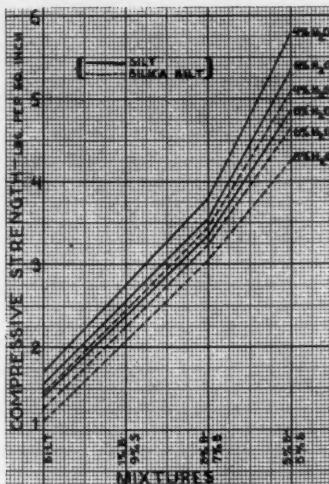


FIG. 8.—MIXTURES OF BANDING SAND, SILT AND CLAY. COMPRESSIVE STRENGTH AS AFFECTED BY MOISTURE CONTENT.

and 6, the effect of the texture of the sand is obvious, it being very much higher with the 20-30 sand.

19. One interesting comparison is to be made between the strength and permeability tests. In the strength test the natural silt always gave higher results than the silica silt. In the permeability tests this held true for the coarser Ottawa sand, but it was the reverse for the banding sand.

20. In Figs. 7 and 8 the compressive strengths are given, arranged in the order of descending silt content, with separate curves for each per cent of moisture used. Here it will be observed that strength decreases as silt increases, but in the case of the 20-30 sand the lines converge, while with the banding sand they remain more nearly parallel.

CONCLUSIONS

21. From these preliminary experiments the following deductions would seem possible:

- (1) An increase in the silt content of a sand appears to lower its compressive and tensile strength.
- (2) A fine-grained sand will stand the addition of more silt than a coarse-grained sand, without serious loss of strength.

(3) Clean silica silt has less bonding power than natural silt, whose grains are apparently covered by a coating of colloidal iron oxide.

(4) In the mixtures used, and within the moisture limits of 4 to 8 per cent, the strength and permeability always dropped with an increase of moisture.

22. It is the intention of the authors to carry out further tests using other types of clay and sand mixtures.

DISCUSSION

H. W. DIETERT: Dr. Ries mentioned the hydrometer method. Is that the Bouyoucas test?

DR. H. RIES: The hydrometer method used was that described by A. Casagrande of the Massachusetts Institute of Technology. It works well and is somewhat more refined than the Bouyoucas method, I believe.

L. H. MARSHALL: Dr. Ries' charts showed that the strength of the sands containing 5 per cent silt and 5 per cent bentonite were higher than those containing 5 per cent bentonite alone. I wonder if that would be true if the optimum strengths were considered.

DR. H. RIES: The mixtures were not tested with less than 4 per cent moisture, and I cannot answer that point as yet.

* U. S. Radiator Corp., Detroit.

† L. H. Marshall Co., Columbus, Ohio.

Application of A.F.A. Sand Testing by the Producer

BY C. M. HARDY,* EVANSVILLE, IND.

Abstract

To secure uniform results in sand control in the foundry, the shipment of the sand from the producer should be uniform. To secure uniformity, the producer must test all shipments for permeability, strength, clay content and grain distribution. Since practically all sand deposits vary from layer to layer, these variations must be taken into account in selection and milling to meet shipping requirements. The author lists tolerances of the various properties which are allowable, and gives representative analyses of various types of sands.

1. The American Foundrymen's Association has gone far into the subject of sand control from the foundryman's point of view; much has been said and written on the subject as applied to foundry practice. Little has been said, however, of the importance of controlling molding sand at its source.

Uniformity Essential in Sand Shipments

2. Unless sand is accurately produced and shipments are uniform, the foundryman will have difficulty in controlling his heaps, and losses are certain to occur in the shop. It is a much simpler matter to control heaps with sand of known uniformity than to attempt control with an unknown, non-uniform product.

3. To secure this uniformity in the sand as shipped, it is necessary that the producer analyze every shipment for permeability, strength, clay content, grain distribution and moisture. It is not enough that one test be made on each car, but frequent tests must be made for permeability and strengths which will show if the desired analysis is being maintained.

* President, Houghland & Hardy, and Hardy Sand Co.

NOTE: This paper was presented before one of the sand sessions at the 1933 Convention of American Foundrymen's Association.

4. A record of all tests should be filed for future reference, of course. The producer should always refer to records of past shipments when filling orders. This is essential because if the foundry, carefully controlling its sand, attempts to use molding sand of non-uniform grade, its time and efforts are largely wasted.

Sand Deposits Show Variations

5. Practically all molding sand deposits are non-uniform when considered as a whole. The grain structure of the deposit may be rather uniform but vary greatly in clay content, strength and permeability; others may even vary in grain structure, especially in pan material.

6. In some cases, it had been noted that, as the pit is extended either farther along the sand structure or into it, the deeper the top soil covering the sand, the greater will be the clay content of the underlying sand itself. The higher levels of such a deposit usually carry a much greater clay content than the underlying stratas, although the grain structure may remain practically the same, while the bottom layers often will have a very low clay content and little strength, in some instances running into the sand's natural silicates, free of clay except for the coating that Nature has caused to adhere to each silica grain. In the above-mentioned cases, of course, permeability increases and strengths decrease as tests proceed farther down the pit.

7. The exact opposite is true of other deposits, the sand near the top containing the low clay and the underlaying stratas increasing in clay and strength with lower permeability farther down. In still other deposits the strata of molding sand is very shallow and almost uniform in analyses from top to bottom at any given place, but varies with the topography of the land in clay content, strengths and permeability.

8. It is these varying conditions, even in adjacent sand structures, that make constant testing necessary. Should the producer neglect this most important work, or fail to overcome or offset these irregularities, the consumer pays for the negligence with mounting shop costs.

Securing Uniformity in Shipments

9. Keeping deposit variations in mind, the producer is obliged, if uniformity in component parts is to be furnished, to work out a method whereby he has a pre-determined analysis of the permeability, clay content, strengths and grain distribution of

the entire deposit, and to maintain a well-ordered chart. As the topography of the land changes during the shipping process, so does the analysis of the pit change. Therefore, it is necessary to have accurate knowledge of the structure at all times, and not alone as to conditions a month or a week ago.

10. Some "soft" molding sands do not require milling but give good results merely by the careful removal of analyzed stratas, working from a pre-determined analysis. Most sands, however, are improved by milling and aeration.

11. The producer is fortunate who has deposits of various light, medium and heavy grades in which each grade may be obtained in different clays, permeabilities and strengths, together with the natural silicates of each sand—and all within a radius enabling the producer to deliver them to one central mill where they may be blended. In such cases, where the analysis of each deposit is known, the producer can bring like grades of different analysis and blend them in various ratio so that the results will meet a prescribed analysis, or where he may put together different types of sand of known analysis to meet certain specifications. The producer's limitations are only circumscribed by the character and kinds of his various deposits.

12. It is certain that if molding sand is to be uniform and produced to meet definite specifications, it is necessary to have some very accurate way to proportion the different sands and natural silicates so as to bring them together simultaneously in the direct ratio each kind of material bears to the whole.

Visual Inspection Unreliable

13. The presenting of molding sand to the foundryman by the old "rule of thumb" method—consisting of taking a well-chosen, nicely tempered sample of sand that has been carefully rubbed through the hands, put in a container and handed to the foundryman who has only the look and feel of the sand to judge by—has largely disappeared. There is no question but that it is an advantage to the foundryman to see the sample of sand being talked about, and that it is also necessary to furnish him with the analysis of that sample.

14. The appearance of molding sand often is very deceiving. For instance, a sand judged by look and feel only may appear to be open and of good permeability, whereas it is not because of its high pan material. Another sand may appear to be tighter and lacking in permeability, when as a matter of fact it is open due

to its uniform grain distribution and low percentage of fines or pan material. This cannot be determined by observation or feel.

15. There is another good reason why sand should be sold by analysis. Many old and reliable concerns in different states and districts have developed sand deposits and by very careful, conscientious and painstaking methods of production have made the sands they produce from these various deposits famous in the foundry world. By custom, these sands, instead of being known generally to the foundry trade by grade number or trade name, often are classified to a great extent by the name of the district in which they are produced. Their merits and analyses have been discussed from time to time both pro and con, and their values have become entrenched in the minds of foundrymen.

16. It is not uncommon, however, for sand to be offered to foundries as being produced from near these well-known districts with the claim that they are the same as such and such a sand, and some special inducement offered to foundrymen to get an order. It is often true that the general appearance of the sand is close to what it is supposed to duplicate, but the methods used in the production are often haphazard and without regard to analysis. Probably no guaranteed analysis is presented, but the inference is made in selling, that the analysis is the same as a certain sand intended to be duplicated, and the inducement may be such that an order is given.

17. However, when the car arrives and the sand is used, the foundryman often finds himself the victim of a shrewd salesman. Had the material been bought on a guaranteed analysis, that loss could have been saved.

Advantages of Buying by Analysis

18. Some may ask "What is the advantage of buying by analysis when we do not have testing equipment in our foundry?" The A.F.A. terms and analyses are the grounds on which producer and foundryman meet understandingly. There are very few foundrymen today who are not familiar with the A.F.A. terms of permeability, strength, clay content, grain distribution, pan material and moisture content, and their direct relation to castings. Foundrymen have a definite idea of what they want—they are discussing their difficulties and exchanging ideas on one phase or another of their problems, and reading the many bulletins and recommendations of the A.F.A.

19. Consequently, in ordering from any producer it is to the

advantage of the foundryman that he be familiar with A.F.A. testing and specifications.* In turn, he rightly expects the producer to have a comprehensive understanding of these terms and their application to his molding sand problems.

20. In filling an order, the producer should bear in mind that a molding sand produced in the Albany or Eastern district, of a given clay content, permeability, strength and grain structure, will give a different result from a sand of the same analysis originating in the Middlewest or South. In other words, three different results in the foundry will be noticed on sands of the same A.F.A. analysis from these widely separated districts. All producers should understand these conditions and know what compensating change in the analysis of their own sand is required to give the same service as a sand from a different district. Very probably the foundryman ordering is not acquainted with this fact, but the producer should advise him of any suggested change in analysis and furnish that analysis for the buyer's records.

TOLERANCES IN SPECIFICATIONS

21. A wide shipping tolerance in specifications should not be necessary for the producer who has a careful and accurate control of his sand and adequate facilities for loading, milling, blending and proportioning it. The closer shipments can be held to given specifications that have proven successful in actual foundry practice, the less grief there will be in the foundry.

22. The tolerances discussed below are based on sand being tested at a given moisture. A moisture content of $6\frac{1}{2}$ to $7\frac{1}{2}$ per cent is considered by many to be a good moisture content from which to work. When moisture content varies, permeability and strengths also vary.

Clay Content.

23. A producer should not consider it unreasonable to be required to ship a sand with a specification calling for a clay content within a tolerance or variation from a designated percentage of a total of 3 points, that is, $1\frac{1}{2}$ points either above or below a given specification. To illustrate, let us say the specified clay content on an order was 15 per cent. In this case the clay content in the car should not be lower than $13\frac{1}{2}$ per cent, nor greater than $16\frac{1}{2}$ per cent.

* *Testing and Grading Foundry Sands*, American Foundrymen's Association, 1931.

Green Strength—Light, Medium and Heavy Sands.

24. *Shear.* It should not be difficult to control the green shear strength of light, medium and heavy sands within 0.6 lb. In other words, if a green shear strength specification called for 2.5 lbs., the tolerance should be within 2.2 lbs. to 2.8 lbs., or within 0.3 lb. either above or below the given specification.

25. *Compression.* Likewise, the green compression strength of the same sands should be controllable within a tolerance or variation of 1 lb., or 0.5 lb. above or below a given specification.

26. Table 1 gives examples of sands ranging from the light

Table 1
ANALYSES OF REPRESENTATIVE SANDS* RANGING FROM LIGHT OPEN TO
EXTRA HEAVY TYPES

Example	A	B	C	D	E	F
Type.	Light Open Sand.	Medium Tight Sand.	Medium Open Sand.	Medium Heavy Sand.	Heavy Sand.	Extra Heavy Sand.
Moisture, per cent.....	7.5	7.2	7.0	7.5	7.5	7.3
Permeability	20	18.8	38.2	60	140	500
Green Shear Strength, lbs..	1.3	2.5	2.0	2.8	3.8	2.0
Green Compression Strength, lbs.	6.0	9.0	9.0	10.0	14.8	11.0
Dry Shear Strength, lbs....	5.0	10.0	8.0	8.0	8.0	12.0
Dry Compression Strength, lbs.	25.0	44.0	36.0	44.0	45.5	65.0
Calcium oxide	None	None	None	None	None	—
Fusion Temperature, degs. Fahr.	3092	2840	3038	2786	2894	—
NOTE: "A," sand from pit, being blended or milled. "B," "D" and "E," sands from pit. "C," blended sand. The specimens were screened through common fly screen onto oilcloth which was folded back and forth several times to mix.						
Clay, per cent.....	13.2	17.6	18.0	18.0	18.0	18.0
Per cent Grains Remaining on Screens of Mesh No.						
6	—	—	—	—	—	3.2
12	—	—	—	—	—	10.8
20	—	—	—	—	2.20	23.8
40	—	0.04	1.60	1.60	13.60	31.6
70	0.20	14.20	20.60	40.40	47.00	10.4
100	0.60	17.60	27.60	12.40	10.60	0.8
150	21.80	12.60	11.60	6.00	3.40	0.6
200	58.00	9.40	8.00	5.40	1.80	0.4
270	2.20	7.20	2.80	3.40	0.00	0.2
Pan.....	3.20	20.00	9.60	12.00	2.40	0.2
Total.....	99.20	98.64	99.80	99.20	99.60	100.0

to extra heavy types. The analyses of those sands are representative of the indicated types.

Permeability Control.

27. *Light, Open-Type Sand:* The permeability of a light, open-type sand should be controlled within a tolerance of 2 points. In other words, this classification of sand can be controlled in permeability to within 1 point either above or below a given figure. Example *A* of Table 1 is an analysis of a sand that would be classified as a light, open-type sand.

28. *Medium Tight Sand:* The permeability of a medium tight sand (Example *B*, Table 1) should be controllable within a tolerance of 2 points, either 1 point above or below a given specification.

29. *Medium Open Sand:* A medium open sand (Example *C*, Table 1) can be controlled with a tolerance or variation of 4 points in permeability, either 2 points above or below a given specification.

30. *Medium Heavy Sand:* A sand designated as a medium, heavy-type sand (Example *D*, Table 1) should be controlled within 6 points in permeability, 3 points above or below a given specification.

31. *Heavy Sand:* Sand designated as of the heavy type (Example *E*, Table 1) should be controlled within a tolerance of 10 points in permeability, 5 points above or below a given specification.

32. *Extra Heavy Sand:* In sand designated as being extra heavy (Example *F*, Table 1), the tolerance in permeability is more variable and a little more leeway should be granted in the matter of permeability.

Pan Material.

33. Pan material should be controlled within 6 points tolerance, that is, 3 points above or below a given specification.

Grain Fineness.

34. It is expected that the producer will follow the instructions set forth by the A.F.A. for determining grain fineness.

Fusion Point.

35. A very important feature of sand analysis is being overlooked generally by the foundry trade. This feature is the fusion point of molding sand or its related sintering point. The fusion point of molding sand has a great bearing on castings poured at

a high temperature. This subject should have more consideration when selecting molding sand than is generally accorded it at present.

CONCLUSIONS

36. The practice of selling molding sand by sample and without specification and analysis is as out-moded today as the old practice of selling pig iron by fracture. The severe and exacting competition encountered among foundrymen requires that selling methods be on the most comprehensive basis.

37. It is difficult indeed, under the old-fashioned method of selling molding sand merely by sample examination, to put in understandable terms just where a deficiency lies and what is actually desired if a shipment has been found "wanting." The ordering and shipping of sand by analysis—grain fineness, clay content, strengths and permeability—is not an occasion for the foundryman to reject sand that does not come to his exact specifications, but it is an opportunity for the producer and the foundryman to meet on one common, understandable basis.

38. Undoubtedly greater benefit may be secured by the foundryman and the producer of a questioned shipment in the exchange of sand tests than by summary rejection of the shipment. If the use of the sand in the foundry and the production of sand at its source is to be improved, the full cooperation between producer and consumer is paramount, through the medium of tests, analysis and specifications.

Grain Structure Control Insures Mold Permeability Control

By HARRY W. DIETERT,* AND FRANK VALTIER,* DETROIT

Abstract

The shape of sand grains of a molding sand has a decided influence on the permeability and strength of a sand. Permeability of a rounded-grain sand is greater than that of an angular-grain sand, while the strength is less influenced except in cases of core sands where oil is used as a bond. The rounded-grain sands produce a stronger core than angular-grain sands when the same amount of oil is used. The size of sand grains is extremely important, and the authors go into detail as to effects produced by various combinations of base materials. The fine materials below 100 mesh have a great influence in lessening permeability.

INTRODUCTION

1. The art of mixing sands of different size grains to secure a desired permeability is a present-day problem of most practical importance. At times the task of controlling the permeability of the production sand or mold is perplexing.

2. It is well known that the permeability of production sand or of a mold will decrease rapidly with the slightest decrease in fineness, but to increase the permeability by adding a coarser sand is very slow and requires additions in large quantities. Permeability may be increased rapidly by reducing the amount of fines, for example, material retained on sieves 200 and finer. The reason for this behavior should be understood by every foundryman.

3. The behavior of sand mixtures may be better understood by comparing the relation between grain structure and base permeability in place of using mold or green permeability, since base

* Dietert-Detroit Co.

NOTE: This paper was presented and discussed before one of the sand sessions at the 1933 Convention of American Foundrymen's Association.

Table 1
EFFECT OF GRAIN SHAPE ON SAND PROPERTIES

	Rounded Shape.	Angular Shape.
A.F.A. Avge. Fineness.....	50	50
Approx. Surface Area, sq. in. per cu. in. of sand	10,000	15,000
Base Permeability	200	180
3% Moisture, 8.3% Clay:		
Green Permeability	150	120
Dry Permeability	170	140
Green Compression, lbs.....	9.8	11.3
Dry Compression, lbs.....	29.9	39
Green Tensile, oz	16	16.5
Oil-Bonded Core, 1-56 ratio tensile strength-lbs.	268	185

permeability eliminates variations such as may be caused by moisture or clay content.

DEFINITION OF TERMS

4. Before proceeding further, it is desired to establish meanings for some terms which will be used more or less frequently in this paper. These terms are defined below.

Coarse Material.

5. Sand grains which form grain material remaining on sieves above the sieve containing the base grains (A.F.A. average fineness) will be classified as coarse material.

Fine Material.

6. Sand grains which form grain material remaining on sieves below base grains mesh (A.F.A. average fineness) will be classified as fine material.

SHAPE OF SAND GRAINS

7. The question often has arisen as to the effect of the shape of sand grains on the permeability. Various test values, such as grain size, surface area, permeability and strength, are tabulated in Table 1 for rounded (spherical) and angular (sharp) sand grains, when mixed with bonding material by the muller.

8. An interesting fact is that the permeability, whether base, green or dry, is higher in each case by at least 20 per cent for the rounded sand grains than for the angular. The above fact is

most natural, as rounded grains produce maximum void space.

9. The strength values of rounded grains when bonded with clay produce a molding sand with slightly less bond strength than a molding sand made with angular sand grains. Clay bond does not spread entirely over sand grain surfaces as an oil film; thus, it is not affected by the increase of surface area of the angular grains.

10. When these same sand grains are bonded with an oil core compound and baked, the core made with the rounded grains will have a much higher strength than the angular sand core. The larger surface area of the angular grains demands a greater quantity of oil, which accounts for its lower baked strength.

SIZE OF SAND GRAINS

11. The relation between sand grain size and base permeability is worthy of study to obtain a clear conception of the permeability power of each sand grain size.

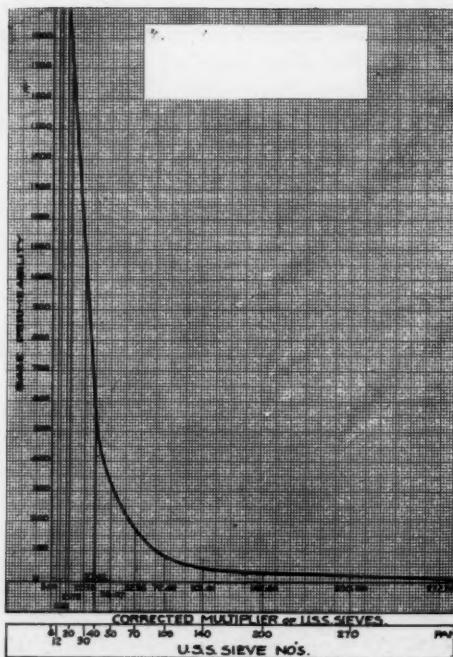


FIG. 1—BASE PERMEABILITY CURVE OF SAND GRAINS ACCORDING TO MESH SIZE.

Table 2
LABORATORY BASE PERMEABILITY, AND BASE PERMEABILITY
CALCULATED FROM FORMULA A (PARAGRAPH 15)

Retained on Sieve.	Multiplier	Laboratory Base Permeability.	Calculated Base Permeability.	Difference per cent.
Pan	270	6.8	6.7	- 1.47
270	200	14.0	12.3	- 12.1
200	142.86	27.0	24.0	- 11.1
140	101.01	45	46.7	+ 3.8
100	72.46	86	94.5	+ 9.9
70	52.36	196	167	- 14.8
50	38.02	313	303	- 3.2
40	27.62	535	540	+ 0.93
30	20.16	950	960	+ 1.05
20	10.72	1796
12	5.81	6075	6500	+ 6.9

12. The base permeability for the different sand grain sizes is plotted in Fig. 1. The ordinate of this chart is shown both in correct sieve multipliers and in sieve sizes. The base permeability for the various sand grains is also tabulated in Table 2.

13. The size of sand grains as found in molding sands in general use may range from 20 mesh to pan material. The 20-mesh material possesses a permeability of 1796, while pan material has a permeability of 6.8. The diameter of coarse material is 15.5 times greater than that of the pan material, while the permeability is 283 times as great.

14. The point of interest is that the permeability of the sand grains is not proportional to diameter, but is a parabolic function. Stated in practical terms, the permeability of the sand grains increases very slowly as sand grains increase from pan material up to 140 mesh, and then it increases very rapidly as sand grains increase in size above 140 mesh.

15. A formula to compute the base permeability of a given grain size is as follows:

$$\text{Base Permeability} = S \left\{ \frac{714}{3+M} \right\}^2 \quad \text{Formula (A)}$$

Where S = Grain sharpness.

Round = 1.0; Angular = 0.8

M = A.F.A. correct multiplier.

16. The tabulation of Table 2 will be of interest to those who are interested in checking laboratory tests with base permeability as calculated by *Formula (A)*.

17. A comparison of the calculated and laboratory base permeability figures shows an agreement, considering the fact that it

is very difficult to secure rounded sand grains throughout the whole range. The formula undoubtedly is of sufficient accuracy for practical purposes. However, extensive research work is required to find a means of measuring sand grain sharpness.

PERMEABILITY CONTROL

18. The increasing or decreasing of permeability is best studied by mixing two different sizes of sand grains at the same time. Sand grains coarser and finer than 70 mesh may be mixed with 70-mesh sand grains in various percentages to show why permeability of production molding sand decreases so quickly and increases so slowly.

19. Base permeability curves illustrating the permeability change for either "Close up" or "Open up" for 70-mesh sand are shown in Fig. 2.

Close Up.

20. Consider first the "Close up" reaction. Add either pan or 270-mesh material to 70-mesh sand, and it may be noted from

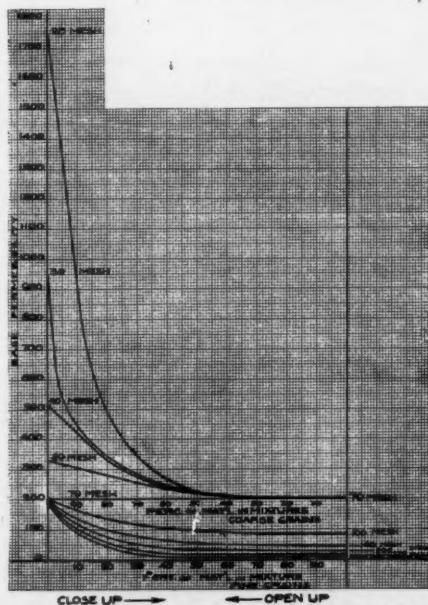


FIG. 2—BASE PERMEABILITY OF SAND GRAINS CONSISTING OF A.F.A. STANDARD SAND AND ALL OTHER SIEVE SIZES.

Fig. 2 that a 10 per cent addition will cause the permeability to drop from 200 to 100.

21. The permeability of a sand drops very rapidly with the slightest addition of fine material, as illustrated by the steep slope of the left-hand portion of curves in Fig. 2. This illustrates how quickly sand closes up.

22. It may be stated that the fine sand grains will have absolute control of the permeability of the sand when they constitute 70 per cent by weight, the reason being that the fine material will lie in the void space between the large sand grains and reduce the gas passageways.

23. A 70 per cent addition of a finer sand grain material will cause the permeability of the sand to reduce almost to the permeability of the fine material added. In a sand of this nature, one would have 70 per cent fine material and 30 per cent of coarse material, with the resultant permeability equaling the permeability of the fine material; for example, 70 per cent of 140 mesh and 30 per cent of 70 mesh, or 70 per cent of 70 mesh and 30 per cent of 20 mesh.

Open Up.

24. The "Open up" process is the reverse of the "close up," working not on the steep-slope side of the permeability curves but on the flat side, namely, the right-hand side of the curves of Fig. 2.

25. Assume that production sand is 70-mesh material; add 10 per cent of 20-mesh material to "open up." The base permeability of the sand will not increase, and the green permeability also would remain the same providing the percentage of clay bond and moisture remains constant. Additions up to 30 per cent will have but slight effect in increasing permeability.

26. The reverse of the "close up" rule applies in the "open up" process in that the production sand becomes in the latter case the fine material. As long as 70 per cent of the original production sand is present, it will control the permeability by filling up the large void spaces as may be formed by the coarse material added.

27. When the percentage of coarse material added increases above 30 per cent, it then begins to build sand structures free, or partly free, of fines.

28. The fine grains of a sand have a greater influence on the permeability of the sand than do the coarser grains, inasmuch as they are free to fill void spaces, while the coarser grains are sand

structure producers. Thus, it may be stated that the fine grains have greater permeability power or influence than coarser grains.

29. A practical question often arises concerning whether a real coarse sand or a sand more nearly the size of production sand will prove most effective in opening up a sand. Studying the curves of Fig. 2 indicates that the permeability may be increased as effectively by sand of adjacent sieve sizes as by a sand of a real coarse size, when sand additions do not exceed 50 per cent. Above this addition, the coarse size is most effective.

PERMEABILITY OF A THREE ADJACENT SIZE MIX

30. The permeability of single and two-sizes mixes is very slow to change when fine material exceeds 70 per cent. When sand grains from three adjacent sieves are mixed together (see Fig. 3), then 33.3 per cent of each—the fine, coarse and base sand grains—is present.

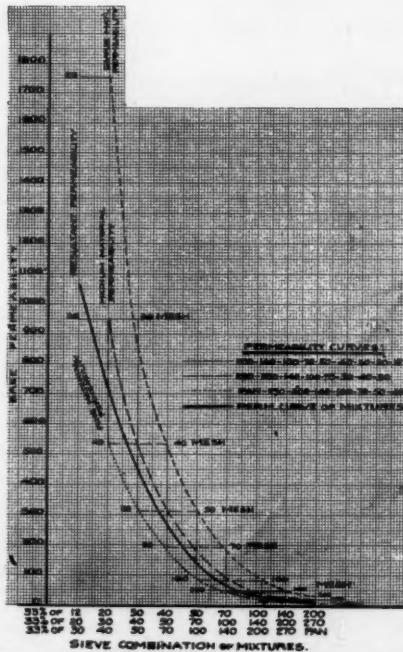


FIG. 3—BASE PERMEABILITY OF SAND GRAINS CONSISTING OF SAND RETAINED ON 3 ADJACENT SIEVES.

31. An insufficient amount of fine material is present to control the permeability, but since the permeability power of fine material is greater than that of the coarse, the resultant permeability of the mix will not be an average of the three sand grains. It will be at least 50 per cent greater than the base permeability of the fine sand grain size.

32. A formula which may be used to calculate the base permeability of three equal adjacent-sieve size mixtures is as follows:

$$\text{Base Permeability} = S - 0.66 \left\{ \frac{714}{3+M} \right\}^2 \quad \text{Formula (B)}$$

Where S = Grain sharpness

M = A.F.A. multiplier of base material or
A.F.A. fineness of mix.

PERMEABILITY OF A THREE NON-ADJACENT SIZE MIX

33. Sand mixtures which consist of three scattered-sieve size sand grains are ideal to show how the permeability is decreased by the scatter or distribution of sand grains.

34. The permeability curves shown in Fig. 4 give the per-

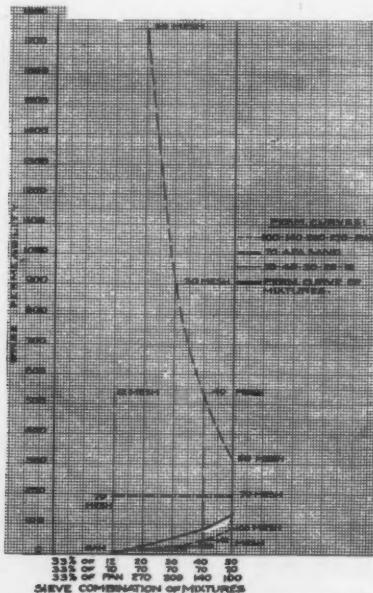


FIG. 4—BASE PERMEABILITY OF SAND GRAINS OF A GIVEN SIZE MIXED WITH COARSER AND FINER GRAINS.

meability of sand containing 33.3 per cent of 70-mesh sand grains as base material, to which equal portions of coarser and finer grains have been added.

35. A sand built up of 50-70-100 mesh material has a permeability of 133. The sand grains are distributed equally on three adjacent sieves. Increase the distribution by omitting one sieve size on each side of the 70 mesh. The sand then would consist of 40-70-140 mesh material in equal portions, with a permeability of 84. The permeability of this wider distributed sand is 37 per cent less than the sand made up of adjacent sieve sizes.

36. The permeability continues to drop as the 70-mesh sand is mixed with sand grains of still greater coarseness and fineness. The drop of permeability is so great that when 12-mesh and pan material are added to 70-mesh material the resultant permeability is 6.0, which is slightly less than that of pan material.

37. The base permeability of the above-mentioned sand grains, 12, 70 and pan, is 6075, 196 and 6.8, respectively. It is a rather forceful point to realize that the final resultant base permeability is 6.0, showing that the 6075 permeability of 12 mesh had no permeability increasing effect on the final base permeability of the mix.

38. When grains of a sand remain on widely scattered sieves, the coarse material, when not exceeding 33 per cent, will not increase the permeability of the sand.

39. Many natural molding sands have a fineness where four sieves carry practically all of the grain material. With this in view, it is of great interest to know the permeability of four adjacent size mixes.

40. The resultant base permeability and green permeability of sand mixes consisting of four adjacent sizes of grains is shown in Fig. 5. Note that the base permeability of a mixture of 70, 100, 140 and 200 mesh grains with base permeabilities of, respectively, 196, 86, 45 and 27, results in a sand with a base permeability of 37.

41. A further study will show that the resultant base permeability lies between that of the two fine-grain materials. As in the case of the three adjacent size mix, the fine material in the four adjacent mix exerts a very strong influence on the resultant base permeability, which will not exceed the base permeability of the second finest material.

42. The formula as shown below may be used to calculate the base permeability of the four adjacent sieve size mixtures:

$$\text{Base Permeability} = 0.50 S \left\{ \frac{714}{3 + M} \right\}^2 \text{ Formula (C)}$$

where S = Grain sharpness, and
 M = A.F.A. fineness number of the mix.

43. The green permeability of a four adjacent sieve size mix will be determined by the base permeability subjected to alterations due to variation of clay and moisture content of the mix.

44. Where these variables, namely, clay and moisture, are held within limits suitable for molding, the relation between base and green permeability as shown in Fig. 5 may be taken as characteristic.

45. The base permeability of a four adjacent sieve size mix will be higher than that of the finest material, but not greater than that of the second highest.

46. The resultant base permeability of a four adjacent sieve

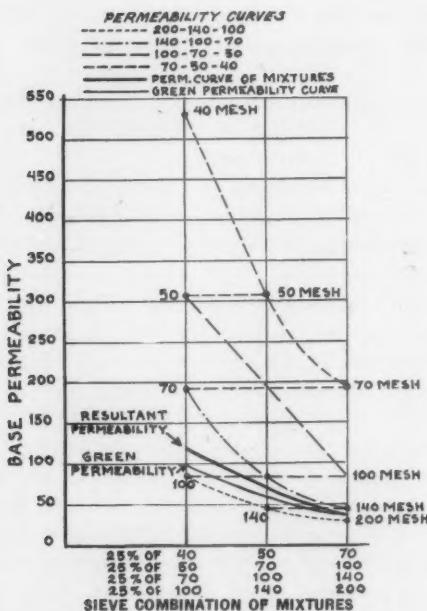


FIG. 5—BASE PERMEABILITY OF SAND GRAINS OF SAND RETAINED ON FOUR ADJACENT SIZE SIEVES.

mix is substantially 0.50 of the permeability of a material corresponding to the A.F.A. fineness number.

SUMMARY

47. Rounded sand grains give at least 20 per cent higher permeability values in both molding sand and baked cores than do angular sand grains, assuming other conditions to be identical.

48. The baked strength of oil-bonded cores made from spherical sand grains is greater than cores made from angular sand grains.

49. The green and dry strength of molding sand is not affected within practical limits by the shape of the sand grains.

50. The base permeability of various grain sizes does not change rapidly until grain size increases above 140 mesh, above which size each increase of sieve size will give an increasingly greater base permeability.

51. A 10 per cent addition of a much finer size sand grain than the base material may cause permeability of production sand to drop 50 per cent. This explains why green or mold permeability drops or "closes up" so readily in practice.

52. A 70 per cent addition of a finer sand grain material than the base material will cause base permeability of production sand to drop very close to base permeability of the fine material added.

53. The permeability power of fine material is large and is sufficient to control the permeability of a sand entirely, as long as 30 per cent or more of fine material is present.

54. A production sand will "open up" (increase in green or mold permeability) when accumulated percentage of coarse grain material exceeds 30 per cent. This explains why it is so difficult to "open up" a sand.

55. Fine sand grains have a greater influence on the permeability of a sand than do the coarse sand grains, and thus have a greater permeability power.

56. When three sand grains from three adjacent sieves are mixed in equal portions, the resultant base permeability will not be the average of the base permeability values for the various sand grains mixed, but much lower, approximately 66 per cent of the average base permeability values.

57. As the scatter or distribution increases in spread, then the base permeability of a sand reduces materially. When spread

is at its maximum, then the resultant base permeability is approximately equal to or lower than the fine material.

58. When the coarse material is not of an adjacent sieve size to base material and the percentage of coarse material is less than 30 per cent, then the coarse material will not help to increase the base permeability.

DISCUSSION

J. T. MACKENZIE: * We have a rather unusual sand, from the ordinary foundry standpoint, in that the grain size runs chiefly between a 12 mesh and a 40 mesh, or perhaps 50, and we use this sand over again every three or four hours. We start with a washed sand base which fact made us more or less interested in Mr. Dietert's discussion of base permeability of green sand. To keep our sand up to condition, we add a clay wash, using a rather fat, yellow clay which we have locally adjacent.

We found long ago that the way to control the permeability of our sand was through the elimination of the burnt clay, and that we get very little effect from adding any quantity, one might say, of the original poor sand. We eliminate the burnt clay in our centrifugal casting molds by keeping a strong suction of air from our dust-collecting machinery over all the shakeout operation or wherever the sand passes from one container to the other. The sand is taken from the shakeout to the treating machinery by a reciprocator which also stirs it up and there is very little tendency for the clay that is still good to come out in the process, although the burnt clay comes out fairly easily and we remove quite a quantity of dust. We require, then, only a fairly small addition of clay to retemper the sand.

As all local clay banks are non-uniform, we find that our permeability reacts very readily to variations in the percentage of clay. We check the moisture content and clay content, of course, and control the sand through the two properties of strength and permeability. Then, if we have a poor grade of clay, we find our permeability dropping down to the limit of, say, 350, whereas with good clay we can readily keep a permeability of 500. Our sand is a beautiful illustration of the rapidity with which that permeability varies. We have a very safe working range of permeability, one might say, for our castings—from 400 to 500, a 100-point range; in case of a very good streak of clay, we will operate on a permeability of 700.

This elimination of dust is a problem having a great many angles right now, an important one being the legal angle. It appears to me that we have not sufficiently emphasized the value of dust collection with the

* American Cast Iron Pipe Co., Birmingham, Ala.

agitation on the shakeout from the standpoint that Mr. Dietert has brought out.

There is also another very good way to eliminate dust where the air suction cannot be used, and that is in a complete washing of a certain percentage of sand each time. We took up with the Dorr Company the matter of following this method as an auxiliary scheme. My recollection is that it would not be very expensive to wash completely, say, some 15 or 20 per cent of the sand by means of water and a Dorr classifier unit. This apparatus would not take up very much room, especially for coarse-grained sand. Of course, the finer the sand, the more trouble it would be; but our shop experience bears out Mr. Dietert's findings.

DR. H. RIES: * In looking over the list of papers given at our meetings since the work on sand research was first initiated—about ten years ago—it is interesting to see how smaller and smaller details are being given attention. Thus, Mr. Dietert has referred to the shape and the surface character of sand grains and their effect, a thing that we did not think of ten years ago.

With regard to that one point of shape of sand grains in comparing the rounded with the angular ones, that, of course would apply only to the coarser sizes, because it is practically impossible to find round sand grains on the smaller meshes.

Mr. Dietert has brought out a number of interesting points in his paper, and when I saw the preprint of the paper it reminded me of some very interesting data given me about a year ago by C. Mathiesen of Whitehead Bros. Co. with permission to publish them in the committee reports. Owing to the fact that the committee has not made formal report lately (or at least has not published any), these figures have remained unpublished. What Mr. Mathiesen did was to separate the different sizes of sand grains from a number of different sands; he then made up combinations of equal amounts of the different meshes and determined the base permeabilities of the mixtures. These data are presented here as Figs. 6 and 7 and Table 3, and they bear out exactly the points made by Mr. Dietert.

I want to ask Mr. Dietert two questions. I believe he said, in paragraph 36, that when he made a mixture of 12 mesh, 70 mesh and pan material, he got a permeability of practically zero. How much would the addition of the clay raise that permeability?

The other question is this: Does the type of clay that is used to mix with these sand grains make a difference in the permeability obtained? In other words, if 5 per cent of one kind of clay is added to the mixture and then 5 per cent of another kind of clay to the same mixture of grains, will different degrees of permeability be obtained?

H. W. DIETERT: In answer to your second question, Dr. Ries, I think you certainly would get different green permeabilities by using different clays or bonding materials, and different percentages. Up to a certain percentage, the clay material will pick up the fines. After that, the clay material begins to fill up voids, just the same as do fine materials, and if you would draw a curve showing the relation between the percentage

* Cornell University, Ithaca, N. Y.

of clay in the sand and the permeability, you would find it was also a parabola. At the beginning, the permeability would not be affected greatly by the low clay content, but as soon as the clay content increased, the permeability would quickly drop.

With reference to your first question, as to the effect of the addition of clay to this wide distribution, if we take the 50, 70, 100 and 140 mesh material and add those together and mix them up well, our base permeability would be, using no clay, 75. The green permeability, adding the 8.3 per cent of clay, which will give a sand of the correct strength for molding, would drop to 60.

In other words, the addition of clay did not increase the green permeability in this case, but I believe that if we had added bentonite, our green permeability would have been higher than the base permeability. There are cases where the green permeability is higher than the base permeability.

Table 3
BASE PERMEABILITIES OF ALBANY MOLDING SAND
(C. MATHIESEN)

(Upper line indicates sieve sizes. Lower lines show percentage of grains of each size which give the base permeability shown in last column.)

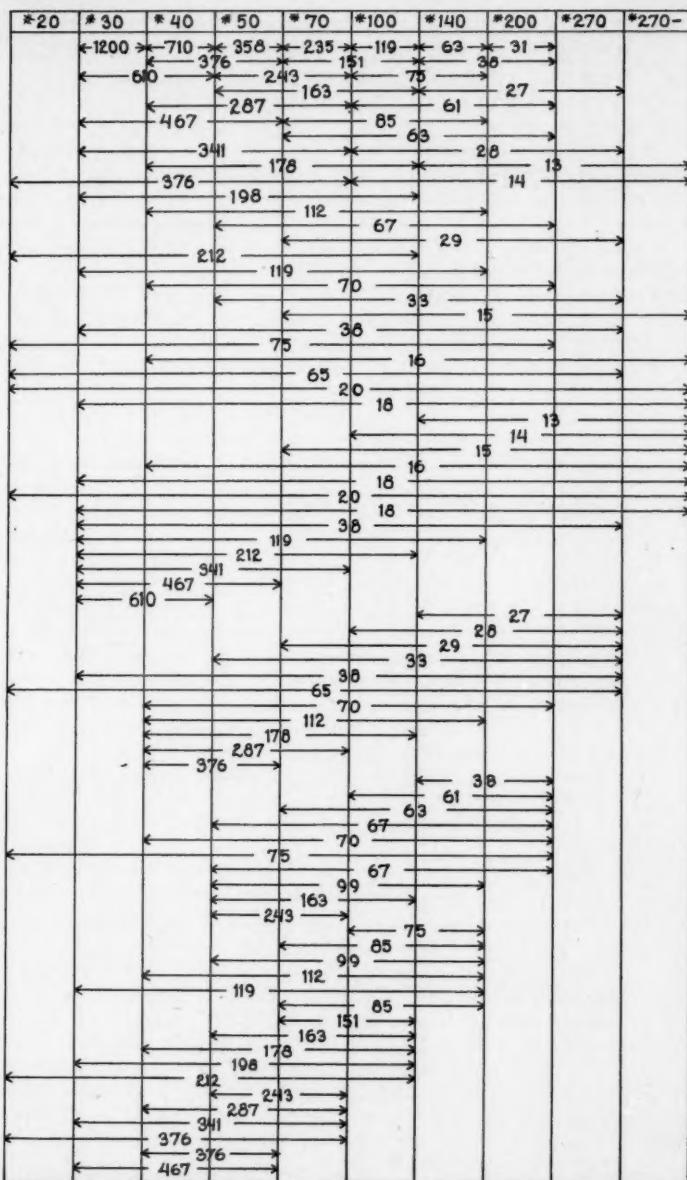


FIG. 6—BASE PERMEABILITIES OF WAREHAM, MASS., SAND (C. MATHIESEN).
SIEVE SIZES ARE SHOWN AT TOP OF EACH COLUMN. HORIZONTAL LINES GIVE PERMEABILITIES FOR VARIOUS MIXTURES, ALL MIXTURES CONSISTING OF EQUAL PORTIONS OF SANDS OF GRAIN SIZES INDICATED.

GRAIN STRUCTURE AND MOLD PERMEABILITY CONTROL

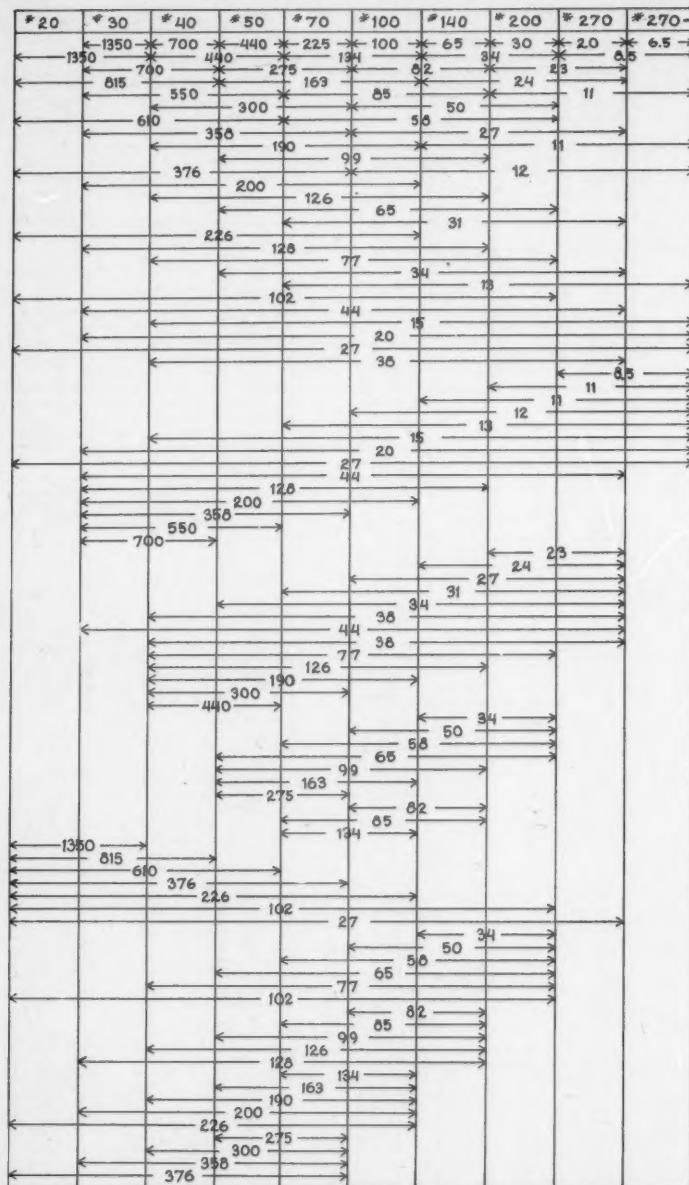


FIG. 7—BASE PERMEABILITIES OF ALBANY MOLDING SAND (C. MATHIESEN).
SIEVE SIZES ARE SHOWN AT TOP OF EACH COLUMN. HORIZONTAL LINES
GIVE PERMEABILITIES FOR VARIOUS MIXTURES, ALL MIXTURES CONSISTING
OF EQUAL PORTIONS OF SANDS OF GRAIN SIZES INDICATED.

DR. H. RIES: What kind of clay was that?

H. W. DIETERT: This was Elfort clay, 8.3 per cent, which would give a compression strength of around 8 lbs., which is a little high for molding. If we had added just sufficient clay to give us about 6 lbs. compression, then the green permeability curve would have been very close to the base permeability. I feel sure that if we had added around 2 per cent bentonite, we would have secured a green permeability higher than the base permeability, due to the fact that the bentonite picked up the finer material and would not close the void spaces of the sand.

Here is another interesting point. We were speaking of fine material consisting of 140. Suppose we go to the beginning of the chart, where the fine material is 200. There the green and the base permeability are approximately the same. If we had extended this chart and begun with pan material, I do not think there would have been any difference between the base and the green permeability, due to the fact that the pan material would have been carried more easily by the clay.

May I add a caution at this point? We are talking about removing the fines, but that is not the whole story. The story is to control the fines.

S. H. CLELAND: * I notice, in the summary of the paper, that the green and dry strength of molding sand is not affected, within practical limits, by the shape of the sand grains. I can remember very well a test we made, looking for information on that point, in which we took an unsatisfactory silica sand, rounded grains, of coarse size, with no strength that could be determined on a strength machine, and we crushed the grains (I cannot say to what grade of fineness, but it was very fine), added moisture to it, and we got a green strength that would be equal to a molding sand green strength. However, it would not make a molding sand because the dry strength would not be sufficient.

The change of the grain size increased the surface area, we reasoned, caused irregularity which developed the interlocking of the sand grains, and with that surface tension and the moisture and the interlocking of the sand grains we got a very decided increase in strength, up to the molding point.

It seems to me that the author's statement about the change in grain size should be qualified by that experience.

There is another point, the question of a decrease in permeability due to the depth of ram. I am wondering if that is true when a sandslinger operation is performed, where the ramming is even from the top to the bottom. Would there be any change of permeability in that case?

H. W. DIETERT: With reference to qualifying the statement on the green and dry strength of molding sand not affected within practical limits by the shape of the sand grains, we tried to bring out the conditions obtained in foundry practice, and there we do not have to worry much about the shape of the grain structure, so far as present-day control is concerned. It did influence it, as Dr. Ries has shown in a paper¹

* Eastern Clay Products, Inc., Buffalo.

¹ Ries, H. and Lee, H. V., *Relation Between Shape of Grain and Strength of Sand*. TRANSACTIONS A.F.A. (1931), v. 30, p. 857.

presented before this Association two years ago, where he obtained a higher strength with irregular shaped grains. This we also show in Table 1. This also depends on the type of bond used and how fine is the material used.

In our comparison here, we have maintained the fineness equal and have not decreased the fineness. Decreasing the fineness would increase compression. If tensile tests were used, I believe the results would have been slightly different.

Symposium—Alloys in Cast Steel

Molybdenum in Cast Steel

BY H. W. GILLETT* AND J. L. GREGG,** COLUMBUS, OHIO

1. The behavior of molybdenum in the cast steels is naturally analogous to its action in similar wrought steels, always, of course, keeping in mind the inherent differences which must exist especially as to ductility and impact in any cast metal, mainly due to the absence of "work."

2. In steel of the usual carbon content for steel castings, especially if the manganese be kept at 0.5 or 0.6 per cent—that is, under what might be termed "alloying amounts"—moderate amounts of molybdenum produce little noteworthy increase in the conventional physical properties of the steel when in the annealed condition. In the normalized condition, however, there is (due to the air-hardening effect of molybdenum) a rise of the yield point, an appreciable although less pronounced rise in the tensile strength, and thus an increase in the elastic ratio. With a low temperature draw of such normalized steels, static ductility and resistance to impact are raised without materially altering the strength figure attained in the simple normalizing.

3. Of prime importance in these steels is the very marked increase in their high-temperature strengths. There is ample indication that simple-annealed or normalized-and-drawn carbon-molybdenum cast steels have virtues for high temperature service not yet fully appreciated. Rys[†] shows a large turbine housing of east molybdenum steel employing this property industrially.

Outstanding Properties in Quenched-and-Tempered Condition

4. The most outstanding properties conferred by molybdenum on steel are only noted ordinarily in the quenched-and-

* Director, Battelle Memorial Institute.

** Metallurgist, Battelle Memorial Institute.

† Reference numbers as shown herein correspond with Bibliography at end of paper.

NOTE: This paper was presented and discussed before the steel alloys session at the 1933 Convention of the American Foundrymen's Association. The discussion of this paper will be found beginning on page 256 where the discussion of the four papers presented at this meeting is reproduced.

tempered condition of the steel. Molybdenum decreases the rate at which the steel must be cooled in order to become martensitic; accordingly, it tends—in proper amount and combination—to produce an air-hardening steel or, in lesser amount and with less of the other alloys, one hardenable by a mild air-quench.

5. Since it so greatly slows down the quick cooling rate usually necessary for hardening, the centers of large sections of molybdenum-containing steels will "harden" much more fully than the centers of carbon or molybdenum-free steels. This depth-hardening throughout a large cross-section makes feasible the air quenching of bulky or intricate castings that would not be amenable to air-quenching in the absence of molybdenum.

6. Martensitic molybdenum steels are more resistant to tempering than are molybdenum-free steels. A higher drawing temperature is required to soften them to a given Brinell hardness than is the case with steels of similar strengths and hardness which are molybdenum-free. Hence, casting strains may be eliminated by long high-temperature draws while retaining high hardness and strength.

7. The rate of cooling after the draw is not so important in molybdenum steels as it is in some other alloy-casting steels, because molybdenum is a specific against "temper brittleness." This is a most important and characteristic property conferred by molybdenum, as the behavior under impact test of some heat-treated alloy steel castings has been completely wrecked because of temper brittleness.

Small Amounts Effective in Combination with Other Alloys

8. It requires a considerable content of molybdenum in a straight carbon-molybdenum steel to produce a high-strength product comparable, by the conventional physical tests, to some other fairly low-alloy steels. Since molybdenum costs 80 to 95 cents (the salt 80 cents, ferroalloy 95 cents) per pound, the plain molybdenum steels are not ordinarily used where static figures alone determine their suitability. It happens, however, that many of the inherent properties conferred by molybdenum on steel are shown as the result of the addition of quite small amounts of it, with great static improvement, if it be used in conjunction with other alloy additions.

9. In fact, the modern (quaternary) molybdenum steels contain only about 0.15 to 0.30 per cent molybdenum. With this small amount of molybdenum, such steels approximate the static prop-

erties of much more highly alloyed nickel and/or chromium steels while actually containing perhaps only half as much of those alloying elements, thus attaining all the good properties which such alloys confer upon steel. Hence, the cost of the small molybdenum content may be well repaid by the saving in larger amounts of their other alloys. The use of molybdenum, therefore, may be economical and deserves consideration whenever an alloy steel is required.

Domestic Supply Available

10. From the economic point of view it is worth remembering that the United States is plentifully endowed with molybdenum, and that its production is not a monopoly but is carried on by several groups, of which two especially are large producers. The element, therefore, is a competitive and not a monopolistic commodity, and the possible hazards of having one's products dependent upon an imported metal are absent.

Percentage of Recovery High

11. A further feasibility of using molybdenum lies in the fact that the recovery of the metal from the ferroalloy—calcium molybdate, or molybdic oxide plus lime—is very high. Since molten iron will reduce molybdenum from its oxide, there is practically complete freedom from loss of the metal when remelting gates and risers. The use of molybdenum introduces no new foundry difficulties and has been said² to minimize some of the old ones; for instance, cracking in the mold.

Some Applications of Combination Alloys

12. Molybdenum is compatible with other alloying elements and, as previously stated, reduces the amounts of those elements which are required. Hence, we have the manganese-molybdenum, the chromium-molybdenum, the nickel-molybdenum, and the vanadium-molybdenum steels, or molybdenum steels containing two, three, or even four of other alloying elements. It enables the attainment of combinations of very high mechanical properties which may be sought in quenched and tempered steels. A favorite steel contains nickel, chromium and molybdenum, the amounts of the elements decreasing in that order.

13. As is general with all alloy steels, if we build up a complex steel containing molybdenum we should be careful to hold down the carbon as we add alloys, so as to get maximum ductility

for a given strength. Most of the casting alloy steels containing molybdenum run from 0.20 to 0.40 per cent carbon and seldom go above the higher figure.

14. An interesting and economical steel for casting is manganese-molybdenum, since molybdenum accentuates the alloying effect of the manganese—which latter effect seems to come into play after an amount of some 0.6 per cent manganese is present. The manganese itself is a cheap constituent. The higher manganese tends to good deoxidation and freedom from non-metallic inclusions. Molybdenum has no such reaction, although it certainly has no harmful effect along these lines.

15. There is an increasing tendency to, and utilization^{3, 4, 5} of cast steels with 1 to 2 per cent manganese and 0.20 to 0.30 per cent molybdenum in the normal carbon ranges, and for ordinary uses, as the manganese percentage goes up the carbon should come down. (See references 1, 2, and 3.)

Analyses and Properties

16. The analyses of some cast steels* tested at the Norfolk Navy Yard are given in Table 1, and the properties of these steels after heat treatment are shown in Table 2. The properties of the manganese-molybdenum steel listed as heat No. 5 compare favorably with those of the other alloy steels, and its high impact value tends to disprove the belief held in some quarters that molybdenum in cast steel is conducive to low impact values. Impact values produced by other heat treatment also were high.

17. An unusual application of higher carbon manganese-molybdenum steel is in car wheels, where a hardness of 300 Brinell is required after a 1575 degs. Fahr. air quench with a subsequent draw high enough to insure freedom from all casting strains. The steel formerly used for their manufacture contained about 0.70 per cent carbon and 1.35 per cent manganese, and was normalized and quenched with a subsequent draw of 1000 degs. Fahr. This draw could not be raised with concurrent retention of the specified hardness. By adding 0.20 to 0.25 per cent molybdenum to the steel, the draw could be raised to 1100 degs. Fahr., giving increased strain relief while still attaining the desired hardness.

18. When cast steel of fine grain is desired, the addition of a little vanadium to a plain molybdenum steel proves a successful

*These steels were made in a $\frac{1}{2}$ -ton, basic-lined, electric furnace. The test bars were cast on the bottom of a slab and removed from the slab with a gas torch. Torch cutting hardened one side of the bars and it was necessary to heat them to about 700 degs. Fahr. to allow machining of the test specimens.

combination. This is utilized at Watertown Arsenal, where centrifugally cast guns replace forged guns in some cases.

Table 1
ANALYSES OF STEELS WHOSE PROPERTIES ARE SHOWN IN TABLE 2

Heat No.	Composition, Per Cent			
	C	Mn	Si	Other elements
2	0.32	1.51	0.29	0.47 Cr, 0.16 V
3	0.16	2.25	0.11	0.11 Cu
4	0.22	2.01	0.09	0.10 Cu
5	0.25	1.12	0.28	0.32 Mo
6	0.38	0.93	0.24	0.36 Mo
7	0.28	0.75	0.21	0.89 Cu
12	0.15	1.43	1.09	0.35 Cu
14	0.30	0.94	0.28	1.45 Ni, 0.17 V

Table 2
PROPERTIES OF CAST STEELS NORMALIZED AT 1700 DEGS. FAHR.,
THEN 1525 DEGS. FAHR., AND FINALLY TEMPERED AT
1300 DEGS. FAHR.

Heat No.	Tensile Strength, lbs. per sq. in.	Yield Point, lbs. per sq. in.	Elongation in 2 in., per cent.	Reduction of Area, per cent.	Brinell Hardness.	Izod Impact, ft-lbs.
2	89,250	70,250	26.5	50.0	160	56
3	113,000	60,750	15.0	36.5	200	14
4	98,750	56,750	25.0	59.0	173	35
5	83,500	64,500	26.0	52.5	150	65
6	81,250	61,250	28.0	55.0		
7	76,500	61,000	31.0	57.0	143	49
12	83,250	57,250	32.0	66.5	145	33
14	88,750	68,500	27.5	55.0	157	31-46

19. Spring found a cast steel of 0.9 per cent chromium and 0.25 per cent molybdenum content to be superior in high-temperature creep resistance to one containing 2 per cent nickel and 0.9 per cent chromium. Again, molybdenum is finding much use in castings containing 5 per cent or more chromium for high-temperature service. The molybdenum addition in these steels is usually about 0.50 per cent; however, a molybdenum content of 0.60 to 0.70 per cent is employed by some producers.

20. At present, cast chromium-molybdenum steels are not so relatively common as are wrought steels of similar compositions. However, their use is extending rapidly on account of their excellent abrasion-resisting qualities. Quenched and tempered nickel-molybdenum cast steels are used extensively and enjoy a reputation for especially good impact resistance for a given strength.

21. It is not necessary to repeat in detail the properties of the above indicated casting steels or of the more complex air-hardening and semi-air-hardening nickel-chromium-molybdenum and manganese-nickel-chromium-molybdenum combinations, since they are tabulated in recent accessible publications by Gregg,³ Lorig and Williams,⁴ and Grotts.⁵ By variations in composition, in quenching and in tempering temperatures, a wide range of properties and many varying combinations of strength and ductility are secured.

22. Heat-treated alloyed steel castings frequently are used in place of forgings for purposes which few today would have thought some years ago could ever be satisfied by castings. Large castings which then would have been made only in the annealed or normalized condition are now being quenched successfully, in air or oil, and subsequently tempered.

23. To sum up, the readiness with which molybdenum fits in with other alloying elements to give steels high in physical properties when quenched and tempered, the added ease and certainty in hardening throughout in quenching produced by molybdenum, and the ability to remove casting strains by a high temperature draw without softening castings too much, make the element a good friend alike to steel foundrymen and to the user of steel castings.

(*For discussion of this paper see page 256*)

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Properties and Uses of Some Cast Nickel Alloy Steels

BY ALBERT G. ZIMA,* NEW YORK

1. Modern industry demands ever greater efficiency and economy of the materials which it utilizes in the construction, operation and maintenance of equipment. The steel foundry industry is among those which have extended their cooperation in meeting these demands, with the result that alloy steel castings now occupy an important position among engineering materials.

2. Among the oldest of alloy cast steels are those containing nickel in amounts from 0.50 to 5.00 per cent. They have maintained an enviable reputation in practically all fields in which they have been applied, due primarily to their excellent static and dynamic properties, their uniformity and reliability. Requirements and typical applications of various types of cast nickel alloy steels are given in Table 1.

3. Nickel produces similar qualitative effects upon the properties of both cast and wrought steels. It enters principally into solid solution in iron (ferrite) and when present in moderate amounts (0.50 to 5.00 per cent) in hypo-eutectoid steels it tends to produce a fine-grained ferrite structure and to refine the structure of the pearlitic areas, thus imparting strength, toughness and, to a lesser degree, hardness without decreasing ductility.^{1†}

4. Excellent dynamic properties, *i.e.*, resistance to impact and fatigue stresses at atmospheric and low temperatures, are obtained as a result of nickel additions. Also, the inhibitive effects of nickel upon grain growth and segregation² are conducive to uniformity of quality of steel castings.

5. As a result of heating above their critical temperatures for extended periods of time, carbon steels suffer an increase in

* Development and Research Department, The International Nickel Co. Inc.

† Reference numbers as shown herein correspond with the Bibliography at end of paper.

NOTE: This paper was presented and discussed before the steel alloys session at the 1933 Convention of the American Foundrymen's Association. The discussion of this paper will be found beginning on page 256 where the discussion of the four papers presented at this session is reproduced.

Table 1*
TYPICAL APPLICATIONS OF CAST NICKEL ALLOY STEEL

Requirements.	Typical Applications.	Type	Composition of Steel.	Section Reference.
HIGHLY DUCTILE STEELS, RESISTANT TO SHOCK AND FATIGUE AT ATMOSPHERIC AND LOW TEMPERATURES.	Locomotive Frames, Castings for Mining, Excavating and Steel Mill Machinery, Ship Castings.	(1) Low-Carbon 2% Nickel Steel, C 0.20 max.; Ni 2.00 min.; Mn 0.60; S 0.15 min.; P and S under 0.05		I
With somewhat higher carbon than (1), increased elastic properties and strength can be obtained.	Miscellaneous Locomotive and Railroad castings, Large Gears not subjected to severe abrasion, Crusher Frames, Tractor and Power Shovel Frames.	(2) Medium-Carbon 2% Nickel Steel, C 0.20-0.30; Ni 2.00-2.25; Mn 0.80-1.00; S 0.15 min.; P and S under 0.05		II
MODERATE COST WITH STRENGTH AND DUCTILITY SUPERIOR TO CARBON STEELS.	Various Construction castings for Tractors, Road-Building Machinery, Etc. Has been used successfully for Electrical Machinery castings up to 6 inch thickness.	(3) Pearlite Nickel-Manganese Steel, C 0.20-0.40; Ni 0.75-2.00; Mn 1.00-1.50		III
These steels characterized by uniform properties. They are easily machined and replace carbon steels when improvements desired in static and dynamic properties do not warrant use of higher alloy steels.		Usual composition, as follows: C 0.25-0.35; Ni 1.00 min.; Mn 1.00 (1.20 max.)		
SUPERIOR MECHANICAL PROPERTIES IN NORMALIZED STEEL CASTINGS. For Light and Medium Sections.	Oil-Well Tools, Sheaves, Sprockets, Tractor Shoes, Gears, Cams, Etc.	(4) Nickel-Chromium Steel, C 0.30-0.40; Ni 1.25-1.50; Cr 0.70-0.90; Mn 0.60-0.90		IV
For Medium and Heavy Sections.	Miscellaneous castings. For Locomotives, Rolling Mill Machinery, highly stressed Gears, Power Shovel castings and other machinery subjected to rugged service.	(5) Nickel-Molybdenum Steel, C 0.30-0.40; Ni 1.25-1.50; Mo 0.20-0.30; Mn 0.60-0.90		VI
QUENCHED AND TEMPERED STEEL CASTINGS WITH HIGH STRENGTH, HARDNESS AND WEAR RESISTANCE. For Light and Medium Sections.		(6) Nickel-Chromium-Molybdenum Steel, C 0.30-0.40; Ni 1.00-2.00; Cr 0.70-0.90; Mo 0.20-0.40; Mn 0.60-0.90		VI
		Nickel-Molybdenum Steel (5).		IV.
		(7) Nickel-Vanadium Steel, C 0.25-0.35; Ni 1.25-1.75; V 0.08-0.15; Mn 0.60-0.90		V
		(8) Nickel-Chromium Steel, C 0.35-0.50; Ni 1.25-1.75; Cr 0.60-0.90; Mn 0.60-0.80		IV
		Nickel-Chromium-Molybdenum Steel, C 0.25-0.35; Ni 1.75-2.25; Cr 0.60-0.90; Mo 0.15-0.25; Mn 0.60-0.80		VI

Table I*—Continued

TYPICAL APPLICATIONS OF CAST NICKEL ALLOY		STEEL—(Continued)
For Medium and Heavy Sections.	Highly Stressed Gears, Pinions, Racks, Rollers, Sprockets and miscellaneous Machine Parts of medium and heavy sections.	Nickel-Chromium-Molybdenum Steel (9). (10) Nickel-Chromium Steel, C 0.35-0.50; Ni 2.00-2.75; Cr 0.75-1.00; Mn 0.60-0.80
ABRASION RESISTANT CASTINGS, For Light and Medium Sections.	Particularly adapted for castings which are to be differentially hardened; for example, Power Shovel Teeth, Impact Hammers, Ball and Tube-Mill Liner Plates, Ore Chute Liners and other parts not subjected to high impact.	Nickel-Chromium Steel (8). Nickel-Chromium-Molybdenum Steel (9). (11) Nickel-Chromium-Molybdenum Steel, C 0.50-0.60; Ni 1.50-2.00; Cr 0.60-0.90; Mo 0.30-0.50; Mn 0.60-0.80
For Medium and Heavy Sections.	Crusher Jaws, Dredge Pumps, Impact Hammers, Dripper Teeth, Etc.	Nickel-Chromium-Molybdenum Steel (12). Steel, C 0.40-0.60; Ni 3.00-3.50; Cr 1.00-1.25; Mo 0.30-0.50; Mn 0.60-0.80
	Resistance to abrasion under heavy pressure and impact. Retention of ductility after exposure to temperatures up to 1100 degs. Fahr. Weldability.	(13) Austenitic Nickel-Manganese Steel, C 0.60-0.80; Ni 3.00-5.00; Mn 10.0-15.0
For Light, Medium or Heavy Sections.	Hot Bed, Run-Out Tables, Forming Dies for Lap-Welded Pipe, Draw Bench Chain for Butt-Weld Tube Mills, Heavy Crusher Rolls, Jaws and Liners, Railroad Crossings.	Nickel-Chromium-Molybdenum Steel (9) with carbon 0.30-0.40
*STRENGTH AND "CREEP" RESISTANCE AT ELEVATED TEMPERATURES.	Valves and Fittings, Return Bends, Superheater and Heat Exchanger Heads, Cement Clinker Chain.	(14) Nickel-Chromium Steel, C 0.30-0.40; Ni 1.75-2.25; Cr 0.70-0.90; Mn 0.60-0.80 Nickel-Chromium-Molybdenum Steel (9) with carbon 0.30-0.40

*Heat and Corrosion Resistant Castings (High Ni, Cr, Fe Alloys) are not discussed in this paper.

grain size and a consequent radical decrease in impact resistance. Nickel steels are not nearly so sensitive to overheating. In the following data, Guillet³ compares the impact properties of a low carbon steel with those of a 2 per cent nickel steel of the same carbon content, after both steels had been subjected to a prolonged heating above their critical temperatures:

Time of Heating at 1830 degs. Fahr.	Ordinary Extra Soft Steel, kg.m.	Extra Soft Steel with 2% Nickel, kg.m.
Normal heating	20.0	60.0 (Not broken)
Four hours	4.5	60.0 (Not broken)
Six hours	4.0	60.0 (Not broken)

6. Nickel obstructs the transformation of austenite to martensite or pearlite, thus causing a lowering of the *Ar* critical ranges. As a result of the retarded transformation period, nickel alloy steels are better adapted for heat-treated castings of large section than are the plain carbon steels. For example, an eutectoid carbon steel must be cooled from 700 degs. Cent. (1292 degs. Fahr.) to 200 degs. Cent. (392 degs. Fahr.) within approximately six seconds if complete hardening and grain refinement are to be obtained. In the case of a 5 per cent nickel steel, this time limit is increased to 60 seconds, while for a 3 per cent nickel, 1.5 per cent chromium steel it is extended to 500 seconds.⁴

7. These differences in critical cooling rates explain why certain nickel alloy steels are particularly well adapted to large sections which cool relatively slowly and in which high strength, toughness and ductility are required. Also, since certain of these steels respond fully to mild quenching in air, the problem of obtaining maximum physical properties in complicated castings is simplified. Many such castings cannot be quenched in oil or water without disastrous results through cracking or distortion.

8. If sufficient nickel is present, the critical ranges are lowered below atmospheric temperatures. Thus, dependent upon the nickel and carbon contents, a nickel steel cooled slowly from a high temperature may be either pearlitic, martensitic or austenitic.

9. The pearlitic group constitutes the greater proportion of commercial nickel steel castings. Compared with pearlitic carbon steels of equal ductility, nickel steels of this group possess higher elastic properties, strength and hardness. Also, the excellent impact and fatigue resistance of pearlitic nickel steels renders them especially adaptable for castings.

10. The martensitic nickel steels are not now of commercial importance because they are not readily machinable, are expensive and relatively brittle.

11. The austenitic steels containing nickel and chromium are rapidly gaining prominence because they provide excellent resistance to heat and corrosion under a wide range of conditions.

SELECTION OF STEELS

12. The selection of a suitable steel for castings is a matter which requires careful consideration. In choosing a steel for a certain application, the engineer considers first of all the properties or performance desired. The size and shape of the piece determine to a great degree the composition and heat treatment to be employed in obtaining these results. Quite often, the type of melting or heat treating equipment available exert an influence upon the selection of a steel. Finally, after these considerations there arises the problem of choosing the most economical steel for the purpose.

APPLICATIONS OF NICKEL ALLOY STEEL CASTINGS

13. Typical cast nickel alloy steel castings, together with some of the uses to which they are successfully applied, are discussed in the following pages. To facilitate a brief but comprehensive presentation, the different steels will be described briefly and correlated with their respective applications.

14. The reader will note that quite often two or more steels may be mentioned as being suitable for some specific casting or class of service. This overlapping may be attributed to several causes. Some of the steels were developed by individual foundries, independent of one another, with the result that often two or more types are used for the same purposes. Again, certain steels possess unique properties which best fit them for certain uses, still they may often be, and are, safely utilized for other purposes as well.

I—Low Carbon 2% Nickel Steel

15. *Typical Applications.* This steel is especially recommended for locomotive frames, castings for mining, excavating and steel mill machinery, ship castings and other parts subjected to shock and fatigue stresses. The retention of impact properties and ductility at low temperatures is another desirable characteristic of this steel, making it well adapted for locomotive and other machinery castings operated in cold climates. (See Table 2.)

16. *Type Composition.* Carbon, 0.20 per cent maximum; nickel, 2.00 per cent minimum; manganese, 0.60 to 0.90 per cent;

Table 2*

NORMALIZED AND DRAWN LOW CARBON 2% NICKEL STEELS—
TYPICAL TEST RESULTS

(Coupons cast and heat treated integrally with castings)

Composition						Physical Properties			Reduction of Area per cent
C	Ni	Mn	Si	P	S	Yield Point lbs. per sq. in.	Tensile Strength lbs. per sq. in.	Elongation per cent	
0.20	2.11	0.81	0.33	0.011	0.024	48,500	81,500	32.0	60.7
0.18	2.16	0.76	0.37	0.015	0.023	48,250	75,750	29.5	61.8
0.20	2.04	0.67	0.30	0.011	0.023	52,500	81,500	32.5	59.9
0.19	2.07	0.89	0.30	0.012	0.024	55,000	81,500	27.5	52.2
0.17	2.09	0.84	0.30	0.012	0.022	51,000	79,000	32.6	62.3
0.19	2.14	0.91	0.30	0.012	0.023	51,250	85,100	29.0	52.0

*Data obtained through courtesy of Canadian Steel Foundries, Montreal, P. Q., Canada.

Table 3

IMPACT TESTS* ON CARBON AND NICKEL ALLOY STEEL CASTINGS
AT ATMOSPHERIC AND LOW TEMPERATURES^b

Composition						Heat Treatment		Temp. of Test, Degs. Fahr.	Impact, ft.-lbs. per sq. in.
C	Mn	Si	P	S	Ni	(A. C.=air cooled)			
<i>Carbon Steel</i>									
0.23	0.66	0.27	0.029	0.046	1600° F. A. C.; Re-heated 1450° F. A. C.	68	245
<i>Nickel Steels</i>									
0.16	0.78	0.35	0.015	0.026	2.36	1750° F. A. C.; Re-heated 1450° F. A. C.	68	702
0.24	0.91	0.46	0.015	0.026	2.21	1750° F. A. C.; Re-heated 1450° F. A. C.	-36	473
0.16	0.88	0.40	0.015	0.031	2.69	1750° F. A. C.; Re-heated 1450° F. A. C.	68	543
0.18	0.97	0.41	0.018	0.031	1.56	0.10	1750° F. A. C.; Re-heated 1450° F. A. C.	-36	603
<i>Nickel-Vanadium Steel</i>									
0.18	0.97	0.41	0.018	0.031	1.56	0.10	1750° F. A. C.; Re-heated 1450° F. A. C.	68	848
								-36	758

*Size of specimen—0.385 x 0.385 in. x 3 in. with 45 degrees V-notch 0.13 in. deep, 1½ in. from one end. A Riehle, pendulum-type impact testing machine was used.

silicon, 0.15% minimum; phosphorus and sulphur under 0.05%.

17. *Recommended Heat Treatment.* At 1750 degs. Fahr., 2 hours per inch of thickness, air cool; at 1500 to 1550 degs. Fahr., 1½ hours per inch of thickness, air cool; at 1100 to 1250 degs. Fahr., draw; cool in air or furnace.18. *Range of Mechanical Properties:*

Elastic limit, lbs. per sq. in.	45,000-55,000
Tensile strength, lbs. per sq. in.	75,000-85,000
Elongation in 2 in., per cent	32-25
Reduction of area, per cent	65-45
Impact (Izod), ft.-lbs.	80-40

19. When given the full recommended heat treatment, the mechanical properties will approximate the maximum results indicated in Table 2, while the lower values may easily be attained by

Table 4*

MEDIUM CARBON 2% NICKEL STEELS—TYPICAL TEST RESULTS
ON COUPONS CAST AND HEAT TREATED INTEGRALLY WITH
CASTINGS

Composition					Physical Properties				Reduction of Area, per cent
C	Ni	Mn	P	S	Yield Point, lbs. per sq. in.	Tensile Strength, lbs. per sq. in.	Elongation per cent		
0.28	1.99	0.97	0.034	0.040	62,300	94,700	26.5	48.4	
0.28	2.16	1.00	0.035	0.045	60,750	97,100	26.5	53.6	
0.27	1.98	0.91	0.033	0.047	57,150	94,350	25.5	47.5	
0.30	1.95	1.05	0.030	0.071	63,550	104,550	25.0	54.8	

Heat treatment: Normalized 1650 degs. Fahr., drawn 1200 degs. Fahr.

*Courtesy of Adirondack Steel Foundries, Watervliet, N. Y.

means of a simple normalizing and drawing treatment. Table 3 presents a comparison of impact test results.

II—Medium Carbon 2% Nickel Steel

20. *Typical Applications.* This steel is well adapted for miscellaneous railroad castings, ship castings, large gears not subjected to severe abrasion, steel mill machinery, crusher frames, tractor and power-shovel frames, and many other parts demanding higher strength and elastic properties than those given by the lower carbon 2 per cent nickel steels previously described.

21. *Type Composition.* Carbon, 0.20 to 0.30 per cent; nickel, 2.00 to 2.25; manganese, 0.80 to 1.00; silicon, 0.25 to 0.40; phosphorus and sulphur, under 0.05.

22. *Heat Treatment.* Normalized at 1650 degs. Fahr., drawn at 1200 degs. Fahr.

23. Range of Physical Properties:

Elastic limit, lbs. per sq. in.....	55,000- 65,000
Tensile strength, lbs. per sq. in.....	90,000-105,000
Elongation in 2 in., per cent.....	28-22
Reduction of area, per cent.....	55-42

24. Table 4 gives data on the properties of some steels under this classification.

III—Pearlitic Nickel-Manganese Steels

25. In ordinary low alloy or pearlitic steels, manganese is added primarily as a scavenger. It reduces the iron oxides present in the bath, forming a reaction product which readily coalesces and rises to the slag. Moreover, manganese is useful in minimizing the bad effects of sulphur in steel.⁶

26. Manganese is ordinarily contained in cast steels in percentages from 0.50 to 0.80 per cent. In finished steel, amounts in

Table 5
TYPICAL PROPERTIES OF SOME CAST PEARLITIC NICKEL-MANGANESE STEELS

Steels.**	Carbon.	Manganese.	Silicon.	Nickel.
A	0.33	1.20	0.38	1.06
B	0.28	0.84	0.35	1.02
C	0.35	1.15	0.76
D	0.34	1.50	0.51	0.77
E	0.33	1.09	1.15
F 1	0.33	1.13	0.39	1.07
F 2	0.32	1.10	0.31	1.10
F 3	0.43	1.38	0.37	1.34
F 4	0.33	1.24	0.38	1.20
G	0.30	1.50	2.00
H	0.34	1.52	0.50	1.30

HEAT TREATMENT AND PROPERTIES

Steels.*	Heat Treatment† (temp. in degs. F.)	Yield Point, Tens. Strength, lbs. per sq. in.	Brinell Hardness	Reduction of Area	Izod Impact, ft.-lbs.
A	1600 F.C.	62,000	100,750	21	34
A	1600 A.C.	63,500	99,000	23.5	187
	1200 D.				
A	1750 A.C.	71,500	106,500	22.8	42.5
	1550 A.C.				
	1000 D.				
A	1750 A.C.	65,000	97,000	25.5	47.5
	1550 A.C.				
	1200 D.				
A	1750 W.Q.	88,000	106,000	21	42.2
	1550 W.Q.				
	1200 D.				
B	1650 A.C.	56,700	87,500	30	45
C	1650 A.C.	48,000	80,000	28	54
	1275 D.				
D	1800 A.C.	65,000	98,500	24	40
	1550 A.C.				
	1280 D.				
E	1650 A.C.	65,000	92,000	25	46
	1250 D.				
F 1	N. and D. to Brinell of about 180	63,650 64,600	92,600 92,300	24.5 25.5	42.7 46.8
F 2	of about 180	63,250	90,400	25.0	40.0
F 3	of about 180	62,450	100,500	24.5	49.2
F 4	of about 180	62,500	96,450	26.0	51.3
G	N. and D.	72,000	101,500	26.0	55.0
H	N. and D.	98,400	121,500	17.0	47.8

*Information sources are designated in 1st column by letters. Thus, F-1, F-2, F-3 and F-4 represent four steels from foundry "F."

**Steel A heat treated in laboratory furnace; remainder of coupons heat treated integrally with castings.

†F. C.=furnace cooled; A. C.=air cooled; D=drawn; W. Q.=water quenched; N.=normalized.

excess of this may be considered to act as an alloy addition, imparting definite properties to the steel. They tend to increase the hardness and strength of the steel.

27. The pearlitic nickel-manganese steels containing from 0.75 to 2 per cent nickel, 1.00 to 1.50 per cent manganese and 0.20 to 0.40 per cent carbon, are rapidly gaining recognition among the low-alloy steel castings. They are moderate in cost, are easily produced by any steel melting process, and possess excellent mechanical properties after simple heat treatments. They are often used to replace carbon and low-alloy steel castings when the improvements desired in static and dynamic properties do not warrant the use of higher alloy steels.

28. Table 5 presents some of the available data on these steels. Fig. 1, plotted from data reported by A. Rys,⁴ shows the

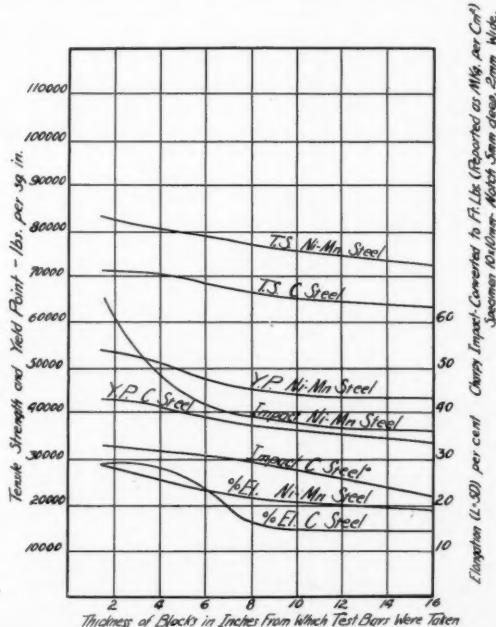


FIG. 1—EFFECT OF SECTION THICKNESS ON PROPERTIES OF CARBON AND PEARLITIC NICKEL-MANGANESE STEEL CASTINGS. (A. RYS) COMPOSITION: CARBON STEEL—0.24 C, 0.40 Si, 0.77 Mn; NICKEL STEEL—0.22 C, 0.17 Si, 0.97 Mn, 0.96 Ni. HEAT TREATMENT: CARBON STEEL—NORMALIZED 1560 DEGS. FAHR., DRAWN 1100-1125 DEGS. FAHR.; NICKEL STEEL—NORMALIZED 1600 DEGS. FAHR., DRAWN 1100-1125 DEGS. FAHR. (SPECIMENS TAKEN FROM CENTERS OF TEST BLOCKS.)

influence of section upon the properties of a carbon steel and a nickel-manganese steel of approximately like carbon content. This latter steel has been used for some time with entire satisfaction for castings up to 6-in. thickness and possibly also in larger sections.

29. *Typical Applications of Pearlitic Nickel-Manganese Steels.* Various structural castings for tractors, motor trucks, road building machinery, electrical machinery and many highly stressed machine parts are made of this steel.

IV—Nickel-Chromium Steels

30. Unlike nickel, which dissolves in the ferrite, chromium combines with the carbon in the steel, forming carbides of high strength and hardness. At the same time it tends to produce fine-

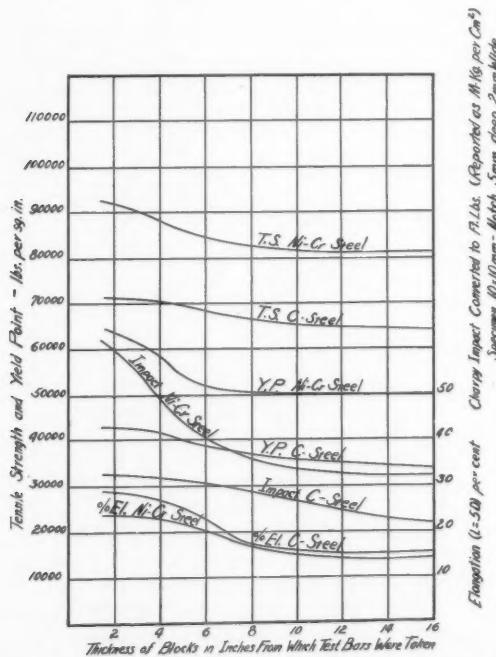


FIG. 2—EFFECT OF SECTION THICKNESS ON PROPERTIES OF CARBON AND NICKEL-CHROMIUM STEEL CASTINGS. (A. RYS) COMPOSITION: CARBON STEEL—0.24 C, 0.40 Si, 0.77 Mn; NICKEL-CHROMIUM STEEL—0.19 C, 0.32 Si, 0.47 Mn, 2.01 Ni, 1.03 Cr. HEAT TREATMENT: NORMALIZED 1600 DEGS. FAHR., DRAWN 1100-1125 DEGS. FAHR. (SPECIMENS TAKEN FROM CENTERS OF TEST BLOCKS.)

ness of structure, thus increasing the static strength of the steel. Through the use of both nickel and chromium, steels may be produced possessing the desirable qualities inherent in each element. The ductility and toughness conferred by the nickel are combined with the strength and hardness imparted by the chromium.⁷

31. These characteristics of the two elements suggest that some ratio of nickel to chromium should be maintained to obtain the best combination of strength and ductility. Experience has shown that a nickel-chromium ratio of 2 or 2½ to 1 constitutes the best balance for the general run of steel castings.

32. Figs. 2 and 3, plotted from data reported by A. Rys,⁴ illustrate the influence of section upon the properties of cast carbon and nickel-chromium steels in the normalized and drawn condition. Table 6 contains data pertaining to normalized and drawn

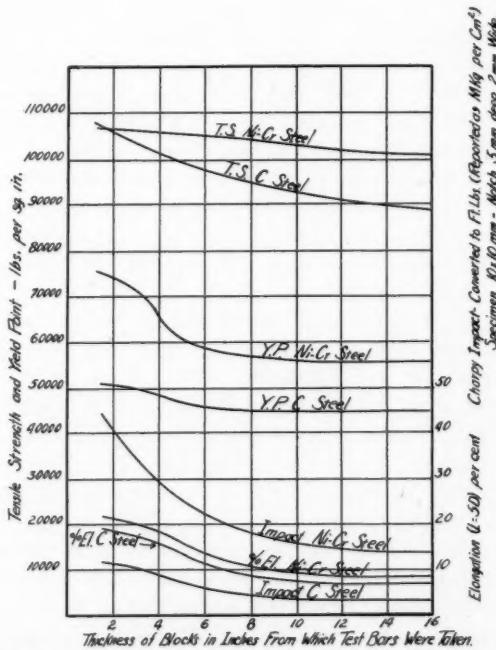


FIG. 3—EFFECT OF SECTION THICKNESS ON PROPERTIES OF CARBON AND NICKEL-CHROMIUM STEEL CASTINGS. (A. RYS) COMPOSITION: CARBON STEEL—0.58 C, 0.32 Si, 0.75 Mn; NICKEL-CHROMIUM STEEL—0.35 C, 0.33 Si, 0.52 Mn, 1.94 Ni, 0.86 Cr. HEAT TREATMENT: NORMALIZED 1560 DEGS. FAHR., DRAWN 1100-1125 DEGS. FAHR. (SPECIMENS TAKEN FROM CENTERS OF TEST BLOCKS.)

Table 6^a

MECHANICAL PROPERTIES OF NORMALIZED AND DRAWN NICKEL-CHROMIUM STEELS*

Chemical Composition					Physical Properties						
C	Mn	Si	Cr	Ni	Tensile Strength, lbs. per sq. in.	Yield Point, lbs. per sq. in.	Elongation, per cent	Reduction of Area, per cent	Izod Impact, ft. lbs.	Brinell Hardness	
0.35	0.80	0.40	0.94	1.30	102,000	67,400	21.0	41.0	38	222	
0.39	0.76	0.46	0.87	1.46	108,500	69,000	19.0	37.0	30	228	
0.37	0.84	0.42	0.92	1.35	104,000	65,300	20.0	35.0	32	217	
0.33	0.75	0.39	0.85	1.50	99,000	60,000	22.5	45.0	41	212	
0.40	0.79	0.38	0.89	1.30	105,200	66,000	18.5	36.0	26	228	
Average					103,000	66,000	20.0	39.0	37	221	
0.37	0.79	0.41	0.89	1.38							

* Type Composition: Carbon 0.30 to 0.40 per cent; chromium 0.75 to 0.90; nickel 1.30 to 1.50. Heat Treatment: Full annealed at 1700 degs. Fahr.; air cooled from 1650 degs. Fahr.; drawn at 1250 degs. Fahr.

nickel chromium steels corresponding approximately to S.A.E. 3140.

33. *Type Composition.* Carbon, 0.30 to 0.40 per cent; chromium, 0.75 to 0.90; nickel, 1.30 to 1.50.

34. *Heat Treatment.* Full annealed at 1700 degs. Fahr.; air cooled from 1650 degs. Fahr.; drawn at 1250 degs. Fahr.

35. Table 6 gives some data on the mechanical properties of some steels under this classification.

36. *Typical Applications of Normalized and Drawn Nickel-Chromium Steel Castings.* Nickel-chromium steels of the composition range indicated in Table 6 are used extensively for oil-well tools, sheaves, sprockets, tractor shoes, gears, cams, and many miscellaneous highly stressed castings.

37. The proportions of nickel and chromium in these steels are so balanced that quenching in oil and tempering emphasizes the strengthening and hardening effects of these elements with maintenance of a high degree of ductility. The retarding effects of nickel and chromium upon the critical cooling rate facilitates hardness penetration on quenching, thus insuring a degree of uniformity of properties throughout the section not obtainable in shallow hardening steels.

38. Table 7 presents data which are representative of the properties of a nickel chromium steel which has been oil quenched and tempered. Figs. 4 and 5 illustrate the effect of section upon oil-quenched and tempered carbon and nickel chromium steels.³

39. Like the chromium steels, some nickel chromium steels exhibit low impact values after slow cooling to room temperature from a high tempering temperature (950 to 1250 degs. Fahr.). This effect, termed "temper brittleness," is not revealed by tensile

Table 7*
REPRESENTATIVE PROPERTIES OF AN OIL QUENCHED AND TEMPERED CAST NICKEL-CHROMIUM STEEL.
Composition: C 0.48, Mn 0.70, Si 0.45, Cr 0.80, Ni 1.24; Size of Coupons 1x1x6 in.
All specimens given a preliminary normalizing treatment at 1650 degs. Fahr, after holding 2 hours at temperature.

Quench Temp. degs. Fahr.	Draw Temp., degs. Fahr.	Yield Point, lbs. per sq. in.	Tens. Str., lbs. per sq. in.	Elong. in 2 in., per cent	Reduction of Area, per cent	Brinell Hardness	Impact (Charpy), ft. lbs.	Endurance Limit, lbs. per sq. in.
O Normalized	1250	65,800	115,750	18.5	28.9	217	35	8.6
A	1450	60,380	114,400	19.5	30.5	217	37	8.3
B	1450	900 Bars Not Machinable	170,350	184,250	8.5	17.0	512	59,000
C	1650	1200	174,000	186,100	7.5	13.4	364	3.1
D	1650	1300	88,460	107,800	24.5	48.6	255	10.9
			89,350	108,000	27.5	47.8	269	50,000
			79,100	103,300	27.0	49.2	197	9.8
			82,650	103,800	26.5	50.0	197	40,000
							35	14.8

*David Zuege, *Alloy Cast Steels*. TRANSACTIONS A. F. A. (1929), v. 37, p. 375.

Table 8
PROPERTIES OF SOME SPECIAL NICKEL-CHROMIUM STEELS

Ref.*	Composition			Heat Treatment††		Physical Properties			Brinell Hardness,	
	C	Ni	Cr	Mn	Si	(Temp. in degs. Fahr.)	Yield Point, lbs. per sq. in.	Tens. Str., lbs. per sq. in.	Reduction of Area, per cent	
A	0.38	2.18	1.00	0.76	0.37	1750 A.C.; 1250 F.C.	95,200	121,300	16.0	32.3
B	0.50	2.80	1.10	0.75	0.30	1650 A.C.; 1300 A.C.	85,000	120,000	25.0	45.0
C	0.40	2.30	1.00	0.82	...	1500 A.C.; 1425 O.Q.	128,000	145,900	20.0	58.6
D	0.35	1.94	0.86	0.52	0.33	1100 D.	78,000	110,000	23.0	...

*Data for steels A and B taken from technical files of International Nickel Co.

†C—Bibliography reference 8. D—Bibliography reference 4.

†A. C.=air cooled; F. C.=furnace cooled; O. Q.=oil quenched; D.=drawn.

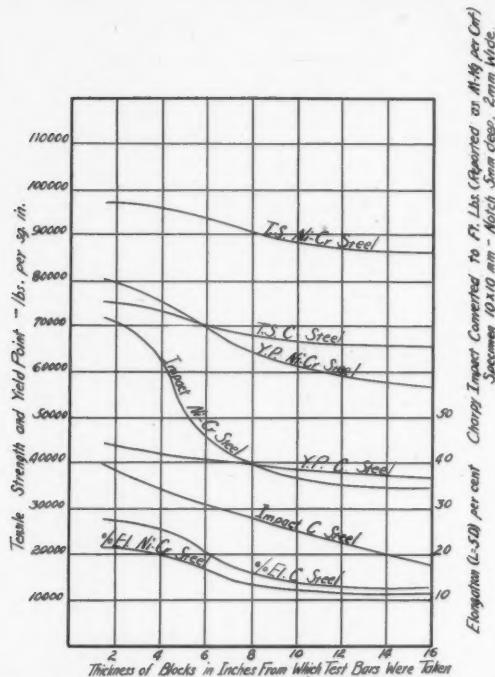


FIG. 4—EFFECT OF SECTION THICKNESS ON PROPERTIES OF CARBON AND NICKEL-CHROMIUM STEEL CASTINGS. (A. RYS) COMPOSITION: CARBON STEEL—0.24 C, 0.40 Si, 0.77 Mn; NICKEL-CHROMIUM STEEL—0.19 C, 0.32 Si, 0.47 Mn, 2.01 Ni, 1.03 Cr. HEAT TREATMENT: OIL QUENCHED 1600 DEGS. FAHR., DRAWN 1100-1125 DEGS. FAHR. (SPECIMENS TAKEN FROM CENTERS OF TEST BLOCKS.)

tests but by the notch toughness of the steel; it can usually be avoided by cooling rapidly from the tempering temperature.

40. *Typical Applications of Oil-Quenched and Tempered Nickel-Chromium Steels.* Oil-quenched and tempered nickel chromium steels containing carbon 0.35 to 0.50 per cent, manganese 0.60 to 0.80, nickel 1.25 to 1.75, chromium 0.60 to 0.90 are extensively used for hardened gears, cams, rollers, sprockets and certain abrasion-resistant castings such as bucket teeth, small crusher jaws, conveyor chain links, etc.

Some Special Nickel-Chromium Steels

41. Cast nickel-chromium steels corresponding approximately to S.A.E. steels 3240 and 3450 are often used for thick-section castings which must withstand high unit stresses. (See reference to

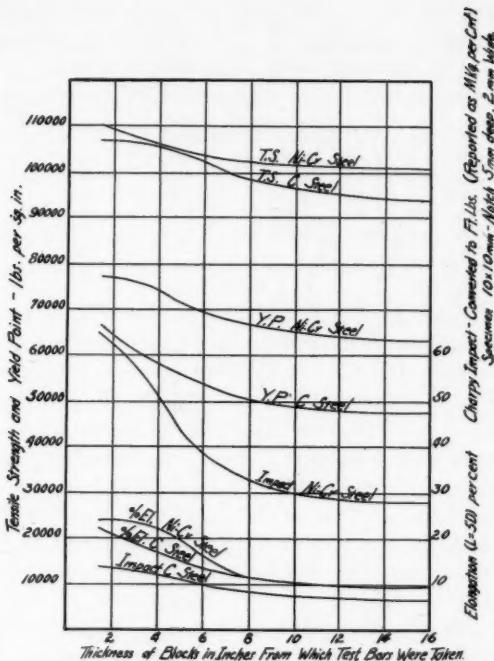


FIG. 5—EFFECT OF SECTION THICKNESS ON PROPERTIES OF CARBON AND NICKEL-CHROMIUM STEEL CASTINGS. (A. RYS) COMPOSITION: CARBON STEEL—0.58 C, 0.32 SI, 0.75 MN; NICKEL-CHROMIUM STEEL—0.35 C, 0.33 SI, 0.52 MN, 1.94 NI, 0.86 CR. HEAT TREATMENT: OIL QUENCHED 1560 DEGS. FAHR., DRAWN 1100-1125 DEGS. FAHR. (SPECIMENS TAKEN FROM CENTER OF TEST BLOCKS.)

critical cooling rates, paragraphs 5-9.) Table 8 presents data on tests made from standard test coupons cast and heat treated integrally with castings, while Fig. 5, plotted from data reported by A. Rys,⁴ illustrates the influence of section upon the mechanical properties of an oil-quenched and tempered 0.58 carbon steel and a nickel chromium steel of the type described in this section.

42. *Applications of Special Nickel Chromium Steels.* Nickel-chromium steels of the type indicated in Table 8 are well suited for many highly stressed gears, pinions, rollers, sprockets and miscellaneous machine parts of medium and large section.

V—Nickel-Vanadium Steels

43. Vanadium, when added to steel in the usual amounts (under about 0.30 per cent), acts as a degasifier, is useful for grain

Table 9*

MECHANICAL PROPERTIES OF CAST NICKEL-VANADIUM STEEL
IN FULL ANNEALED AND DRAWN CONDITION

Composition					Physical Properties				
C	Mn	Si	Ni	V	Tensile Strength, lbs. per sq. in.	Yield Point, lbs. per sq. in.	Elongation in 2 in., per cent	Reduction of Area, per cent	Izod Impact ft. lbs.
0.25	0.94	0.39	1.56	0.11	92,000	61,000	27.0	54.0	60.0
0.20	0.87	0.33	1.50	0.12	85,375	59,000	29.5	59.1	...
0.22	0.74	0.34	1.55	0.10	88,730	56,070	29.5	58.2	63.0
0.26	0.97	0.30	1.54	0.11	90,000	60,500	25.5	50.9	47.0
0.32	1.07	0.28	1.58	0.11	96,500	70,500	28.0	59.9	50.5

*F. Grotts, 8

size control, and also has a tendency to increase the elastic ratio.

44. Vanadium steel castings rarely contain more than 0.15 to 0.25 per cent vanadium. In these proportions a very small part is believed to dissolve in the ferrite, while the major portion is said to combine with the cementite to form complex carbides. Steel castings containing 0.10 to 0.15 per cent vanadium, and 1.50 to 2.50 per cent nickel, exhibit high strength and elastic limit together with good ductility. The grain-refining effect of the vanadium combined with the strength and toughness imparted by the nickel, make this steel especially desirable for highly stressed castings of heavy section.

45. Table 9 presents physical property data from several heats of nickel-vanadium steel in the fully annealed and drawn condition.

46. *Typical Applications of Cast Nickel-Vanadium Steels.* Miscellaneous medium and heavy castings for locomotives and rolling-mill machinery, highly stressed gears and other parts for power shovels and other machinery subjected to rugged service, are produced successfully from this steel.

VI—Nickel-Molybdenum and Nickel-Chromium-Molybdenum Steels

47. Since the World war the use of molybdenum in steels has been developed quite extensively. In order to obtain the full benefits from molybdenum, however, it is generally necessary that it be supplemented by some other alloy such as nickel or chromium, or a combination of nickel and chromium.

48. The nickel-molybdenum and nickel-chromium-molybdenum steels are giving a good account of themselves in castings where high strength and high elastic properties combined with excellent ductility are required. They lend themselves readily to air quench-

ing treatments and are particularly adapted to the production of large castings because of their depth-hardening properties.

49. Also, their ability to retain high strength at elevated temperatures extends their field of usefulness to many industrial applications where heat is involved. (See Table 11.) They are not

Table 10*

SOME TEST RESULTS ON CAST NICKEL MOLYBDENUM STEELS

Composition Range: C 0.30-0.35, Ni 1.25-1.50, Mo 0.30-0.35,
Mn 0.60-0.80Heat Treatment: 1850 degs. Fahr. furnace cooled; 1700 degs.
Fahr. air cooled; 1200-1350 degs. Fahr. draw.(Tests Made from Standard Coupons) Machinability—
excellent

Yield Point, lbs. per sq. in.	Tensile Strength, lbs. per sq. in.	Elongation in 2 in., per cent	Reduction of Area, per cent
60,000	94,000	25.0	55.0
60,750	95,500	26.5	56.8
61,000	94,000	24.0	51.0
62,700	93,860	24.5	53.3
61,000	87,000	23.5	52.0
60,000	86,000	26.0	61.0
60,000	88,000	24.5	51.0
59,000	90,000	24.0	56.0
Avg. 60,536	91,045	24.8	54.5

*These data reported by F. G. Frisbie, Met., Pittsburgh Steel Foundries, Glassport, Pa.

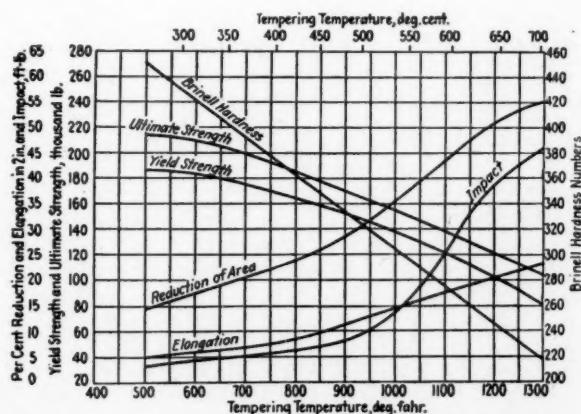


FIG. 6—PROPERTIES OF CAST NICKEL-CHROMIUM-MOLYBDENUM STEEL, QUENCHED IN WATER FROM 840 DEGS. CENT. (1545 DEGS. FAHR.) AND TEMPERED AT INDICATED TEMPERATURES. (BONNEY-FLOYD CO.)

Table 11
COMPARISON OF PROPERTIES OF SEVERAL CAST STEELS TESTED AT
830 DEGS. FAHR. (500 DEGS. CENT.).

Composition	C	Mn	Si	Ni	Mo	Heat Treatment (Temp. in degs. F.) 1690 (920°C.) 8 hrs., furnace cooled	Yield Point, lbs. per sq. in.	Tensile Strength, lbs. per sq. in.	Elong. in 2 in., per cent	Reduction of Area, per cent	Reference.
0.28	0.74	0.37	24,000	25,560	50,126	21.2	28.9	Loring & Williams ⁹
0.35	1.15	0.50	1740 (960°C.) 3 to 4 hours, cooled rapidly to 1025 (550°C.); drawn at 1200 (650°C.)	25,560	50,126	45.0	58.9	Korber & Pomp ¹⁰
0.29	0.92	0.23	2.05	0.28	...	Annealed, 1650-1740 (900-950°C.)	38,000	61,000	12.7	31.8	Liesmann & Salzman ¹¹

Table 12
**MECHANICAL PROPERTIES OF SOME TYPICAL NICKEL CHROMIUM
 MOLYBDENUM CAST STEELS**
*(With Exception of Steel B, Tests Made from
 Standard Coupons)*

^aSteel A reported by A. W. Lorenz, TRANSACTIONS A. F. A. (1922), V. 31, p. 193.
^bSteel B heat treated as coupons 24 in. thick. Reported by J. F. Culp, Vulcan Iron Works.

Barre, Pa.
Data for steel C taken from technical files of International Nickel Co.

difficult to produce, as neither nickel nor molybdenum are oxidized in the bath under normal conditions.

50. Further data on these steels are given in Tables 10, 11 and 12.

51. Fig. 6, obtained through the courtesy of the Bonney Floyd Co., Columbus, Ohio, indicates the properties obtainable with a water-quenched and drawn nickel-chromium-molybdenum steel of the following composition range: Carbon, 0.25 to 0.35 per cent; nickel, 1.75 to 2.25; chromium, 0.65 to 0.90; molybdenum, 0.15 to 0.25. Test bars were taken from 1-in. square coupons.

52. *Typical Applications of Nickel-Molybdenum Cast Steels.* Although the cast nickel-molybdenum steels have not been produced extensively, they have been used successfully for certain highly stressed parts such as rolling-mill machinery castings, heavy-duty gears and pinions of large size, mining machinery castings and various other castings subjected to severe loads. A cast nickel-molybdenum steel containing 0.35 to 0.45 carbon, 0.60 to 0.80 manganese, 0.20 to 0.40 silicon, 0.50 to 1.00 nickel, and 0.20 to 0.30 molybdenum, is used for gears and pinions for steel mill machinery.¹² They are annealed, machined and then heat treated to a Brinell hardness of 380 to 495, depending upon the application. In this condition, these castings possess good wear and shock resistant properties which make them especially adaptable to this severe service.

53. *Typical Applications of Cast Nickel-Chromium-Molybdenum Cast Steels.* The wide range of properties obtainable through variations in heat treatment adapt these steels to many types of service. For example, steel C (Table 12) when annealed is suitable for many structural castings demanding high strength and good ductility. When normalized and drawn (C-2), it may be used for tractor shoes, power-shovel bucket lips, sheaves, sprockets, etc. Gears, sprockets, crane wheels and other castings subjected to wear may be easily machined in the annealed condition and subsequently heat treated to acquire the properties shown for C-3 in the table.

54. The retention of hardness and strength even after exposure to temperatures up to 1000 degs. Fahr. (See Fig. 6) makes this steel especially useful for cement clinker chain, valves, fittings and other castings operating at moderately elevated temperatures. This steel also is used extensively for abrasion-resistant castings.

55. Steel B is well suited for large gears and pinions.

Table 13

(See also Tables 7 and 12)

COMPOSITIONS OF SOME ABRASION-RESISTANT STEEL CASTINGS

Steel No.	Carbon	Manganese	Chromium	Nickel	Molybdenum
H	0.35—0.50	0.60—0.80	0.60—0.90	1.25—1.75
K	0.25—0.35	0.60—0.80	0.60—0.90	1.75—2.25	0.15—0.25
L	0.50—0.60	0.60—0.80	0.60—0.90	1.50—2.00	0.30—0.50
M	0.40—0.60	0.60—0.80	1.00—1.25	3.00—3.50	0.30—0.50
N*	0.60—0.80	10.00—15.00	3.00—5.00

*U. S. Patent No. 1,732,202—John Howe Hall and J. S. Comerford, Assignors to Taylor-Wharton Iron & Steel Co., High Bridge, N. J.

VII—Abrasion-Resistant Castings

56. For a number of applications requiring resistance to abrasion only, chilled cast iron, high-carbon steel or the plain chromium steels are often adequate; but where shock and fatigue stresses are encountered as well, the nickel alloy steels are far superior. The major portion of the alloy steel tonnage devoted to this difficult service is made up of the heat-treated nickel-chromium and nickel-chromium-molybdenum steels. Although the S.A.E. compositions are followed quite closely in steels of this sort, many special steels have been developed. Thus, we find such compositions as are shown in Table 13 giving good service.

57. Steels *H* and *K* of Table 13 are often used for castings which are to be differentially hardened, thus protecting certain portions from excessive wear. Power shovel and scrapper teeth, impact hammers, sprockets, etc., are excellent examples of this type of casting. One method of heat treatment commonly employed for bucket teeth made of nickel-chromium steel *H* is outlined below.

58. The whole casting is first normalized at 1650 degs. Fahr. and drawn at 1250 degs. Fahr. The point or tip is then reheated to about 1550 degs. Fahr. and quenched in oil, after which the whole casting is tempered at 900 degs. Fahr. The properties imparted to the shank and point, respectively, by this treatment are about as follows:

	Yield Point, lbs. per sq. in.	Tensile Strength, lbs. per sq. in.	Elongation, per cent.	Reduction of area, per cent.	Brinell Hardness
Shank	60,000	110,000	20.0	35	212
Point	165,000	185,000	8.0	15	364

59. Steel *K*, Table 13, is similarly employed and is especially useful where high strength and hardness are to be imparted to

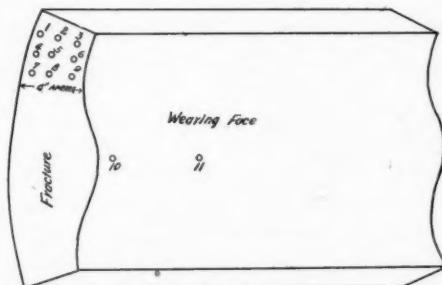
castings of medium or thick section. Properties of this steel are shown under *C*, Table 12. Like steel *H*, it may be differentially hardened to improve the wearing properties of bucket teeth, impact hammers, sprockets, etc. Normalized and drawn, it imparts excellent wear resistance to such castings as power-shovel bucket lips, sheaves, tractor shoes, wearing plates, etc.

60. Steels *L* and *M* are extremely sensitive to air hardening and must be annealed carefully to insure machinability. Instead of allowing the castings to furnace cool from the annealing temperature (1600 to 1650 degs. Fahr.) without interruption as in ordinary annealing, the heat is reapplied at 1200 to 1250 degs. Fahr. and the charge held at this temperature for a short period (1/2 to 1 hour) before furnace cooling is resumed.

61. Steel *L* when normalized at 1650 degs. Fahr. exhibits the following range of properties for test pieces 1 inch in thickness:

Yield Point, lbs. per sq. in.	Tensile Strength, lbs. per sq. in.	Elongation, per cent.	Reduction	
			of area, per cent.	Brinell Hardness
175,000-190,000	200,000-215,000	3 1/2-5 1/2	9 1/2-13 1/2	465-500

62. A low-temperature tempering treatment (600 to 800 degs. Fahr.) minimizes the stresses which are set up on rapid cooling and does not lower the hardness appreciably. In this condition this steel is suitable for ball-mill liner plates, ore chute liners, sand and gravel machinery and other applications not requiring high impact resistance or toughness. Fig. 7, a sketch of a fractured



Location	1	2	3	4	5	6	7	8	9	10	11
Brinell Hardness	415	444	383	448	415	415	383	383	444	383	444

FIG. 7—FRACTURED TUBE-MILL LINER CASTING ILLUSTRATING DEPTH-HARDENING PROPERTIES OF CAST NICKEL-CHROMIUM-MOLYBDENUM STEEL. (COURTESY OF CANADIAN STEEL FOUNDRIES LTD., MONTREAL, CANADA)

tube-mill liner plate casting, illustrates the depth-hardening properties of a steel of this type.

63. Steel *M* is used primarily for wear-resistant castings which, because of their shape or size, cannot be conveniently quenched in a liquid medium. Its excellent depth-hardening capacity makes this steel especially suitable for thick, heavy castings requiring high hardness. Impellers and housings for medium and large size centrifugal pumps handling sand, gravel and other abrasive materials give excellent results when they are made of this steel.

64. These castings are usually air quenched from about 1550 degs. Fahr. after annealing at 1650 to 1750 degs. Fahr., and then drawn at 700 to 800 degs. Fahr. for at least two hours per inch of thickness. Castings 4 to 6 inches in thickness, when heat treated in this manner, acquire a hardness of 380 to 450 Brinell throughout. Usually such castings are rough machined in the annealed state and then ground to finish after hardening.

65. When given the above treatment but drawn at 1200 to 1250 degs. Fahr., this steel is well suited for large gears, pinions, racks, rollers, etc., which are called upon to resist high stresses and abrasion. (See *B*, Table 12.)

VIII—Austenitic Nickel-Manganese Steel

66. Austenitic or high-manganese steel of the Hadfield type is well known as an abrasion-resistant material, but this quality is not exhibited unless the steel is subjected to cold working either before or during service. Heat-treated alloy steels such as *H*, *K*, *L* and *M* of Table 13 are often more serviceable than manganese steel when abrasion without appreciable cold working must be encountered.

67. There are also other factors which limit the use of ordinary Hadfield manganese steel. In order to bring about the austenitic structure required to maintain the toughness of this steel, the castings are heated to 1800 degs. Fahr. or higher and quenched in cold water. This steel possesses a high coefficient of thermal expansion and a very low heat conductivity, a combination of properties which reacts adversely in the quenching of heavy sections. Castings exceeding 5 or 6 in. thickness cannot be quenched without producing internal cracks, which often cause failure in service.

68. Moreover, if heat-treated manganese steel castings are

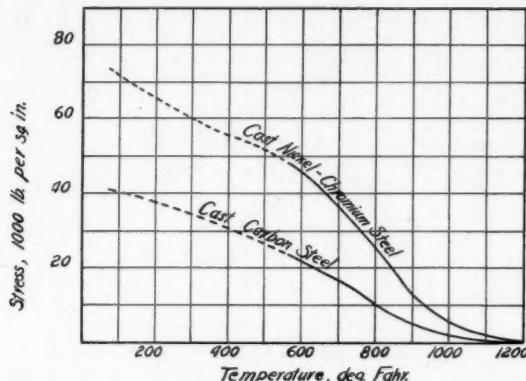


FIG. 8—COMPARISON OF "CREEP" LIMIT STRESSES OF CAST NICKEL-CHROMIUM AND OF CAST CARBON STEEL. (KANTER AND SPRING) THESE CURVES SHOW THE LIMITING STRESSES PRODUCING NOT MORE THAN 1 PER CENT OF CREEP IN A YEAR. DOTTED PORTION CORRESPONDS TO THE PROPORTIONAL LIMIT.

reheated to 600 to 800 degs. Fahr., the dissolved carbides are reprecipitated, thus bringing about extreme brittleness. This phenomenon precludes the use of this steel for service at elevated temperatures and makes welding or forging exceedingly difficult.

69. To overcome these difficulties and to extend the field of usefulness of this steel, a modified composition wherein 3 to 5 per cent nickel is added has been successfully developed. Typical composition, carbon 0.60 to 0.80 per cent, manganese 10.0 to 15.0, nickel 3.00 to 5.00 (See *N*, Table 13).

70. This steel, after a simple normalizing treatment, exhibits the same characteristic toughness and work-hardening properties as ordinary Hadfield manganese steel. Moreover, it retains its strength and toughness even after long exposure to temperatures up to 800 degs. Fahr. or short exposure at 1100 degs. Fahr. This feature permits the use of the steel for such applications as hot bed run-out tables, forming dies for lap-welded pipe, draw bench chain for butt-weld tube mills, and many other applications requiring resistance to wear at elevated temperatures. Large wear-resistant castings for crushing machinery, dredges, etc., are made of this steel since they may be successfully heat treated by normalizing.¹³

XI—Steel Castings for Elevated Temperature Service

71. The nickel-chromium cast steels exhibit excellent creep-resistant properties and strength at elevated temperatures. In

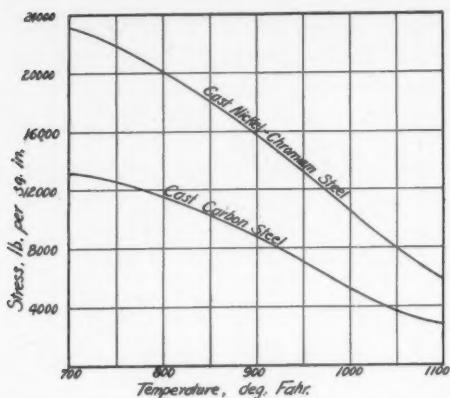


FIG. 9—COMPARISON OF 3 PER CENT "CREEP" LIMIT STRESSES OF CAST NICKEL-CHROMIUM AND OF CAST CARBON STEEL. (MALCOLM) THESE STRESSES WERE DETERMINED ON THE STEELS IN THE AIR-QUENCHED AND DRAWN CONDITION OVER PERIODS OF ABOUT ONE YEAR.

power plants and oil refineries, where materials capable of withstanding high pressures at elevated temperatures are demanded, a steel of the following composition has come into extensive use: Carbon 0.30 to 0.40%, chromium 0.70 to 0.90, nickel 1.75 to 2.25.

72. Fig. 8 compares the 1 per cent creep limit stress values of cast carbon and nickel-chromium steels studied by Kanter and Spring.¹⁴ The compositions of these steels are reported by Bull¹⁵ as follows:

	Carbon	Manganese	Silicon	Nickel	Chromium
	%	%	%	%	%
Carbon Steel	0.33	0.77	0.34
Nickel-Chromium Steel	0.41	0.58	0.29	2.00	0.87

73. Malelom¹⁶ made a similar study in which he compared the limiting stresses producing not more than 3 per cent creep over periods of about one year. The comparison was made with a plain carbon steel containing 0.35 per cent carbon and a nickel-chromium steel containing 0.30 to 0.35 per cent carbon; 0.70 to 0.90 per cent chromium and 1.75 to 2.25 per cent nickel. The steels were tested in the normalized and drawn condition. An examination of Fig. 9 indicates that for temperatures from 750 to 1100 degs. Fahr. the nickel-chromium steel resists creep stresses 70 to 100 per cent greater than those allowable for the carbon steel.

74. Recent investigations have shown that the addition of 0.25 to 0.35 per cent molybdenum to a nickel-chromium steel of

about the same composition, as indicated above, increases the creep resistance appreciably.¹⁷

CONCLUSION

75. The properties and uses of some cast nickel alloy steels have been outlined. Since the scope of the paper does not permit a discussion of all of the nickel alloy steels used for castings, the author has endeavored to limit the text to a brief description of some of the better known compositions and applications. The data presented are not experimental but are representative of the results obtained by many producers.

(For discussion of this paper see page 256)

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¹⁴ Kanter, J. J. and Spring, L. W., "Steels Are Now Available for High Pressures and Temperatures." *Power*, v. 69, May 28, 1929, p. 904.

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¹⁶ Malcolm, V. T., "Nickel-Chromium Steels for High Temperature Service—Valves and Bolts." *Bulletin No. 15, Nickel Steel*, International Nickel Co.

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Vanadium In Steel Castings

By JEROME STRAUSS,* BRIDGEVILLE, PA.

1. Recognition of the benefits of including vanadium in the composition of steel castings is not new. Carbon steel castings were familiar products many years prior to the commercial introduction of the alloying elements; nevertheless, tonnage production of alloy steels appeared first in forgings and rolled bars, although the early use of these elements was by no means confined to hot worked products. Small-scale manufacture of wrought vanadium steel dates back to 1896. It was not, however, until some ten years later, following the discovery of extensive ore deposits in the Peruvian Andes, that continuous manufacture for industrial constructional applications began.

2. Not long thereafter (in 1907, to be precise) vanadium steel castings were being placed in service. Their desirable qualities were quickly recognized so that, to name one specific instance, by the end of 1912 some 6000 locomotive side-frames had been produced. Quite naturally, the initial castings were of carbon-vanadium steel, and this type not only enjoyed continuous and extensive application but remains in wide use today. In recent years, however, the demands of the mechanical industries for construction materials of greater capacity have resulted in newer types involving simultaneous use of several alloying elements.

Influence of Vanadium on the "As Cast" Condition

3. The influence of vanadium upon the structure and properties of cast steels is first observed in the "as cast" condition, not heat treated. Other components of the composition being alike, and the details of steelmaking, pouring, etc., being practically constant, a vanadium-containing steel will exhibit less marked dendritic segregation and greater freedom from Widmanstättian patterns within the grains than a vanadium-free steel.

4. The first-mentioned distinction may be observed in the fracture or in such etched sections as are reproduced in Fig. 1.

* Chief Research Engineer, Vanadium Corp. of America.

NOTE: This paper was presented and discussed before the steel alloys session at the 1933 Convention of the American Foundrymen's Association. The discussion of this paper will be found beginning on page 256 where the discussion of the four papers presented at this session is reproduced.

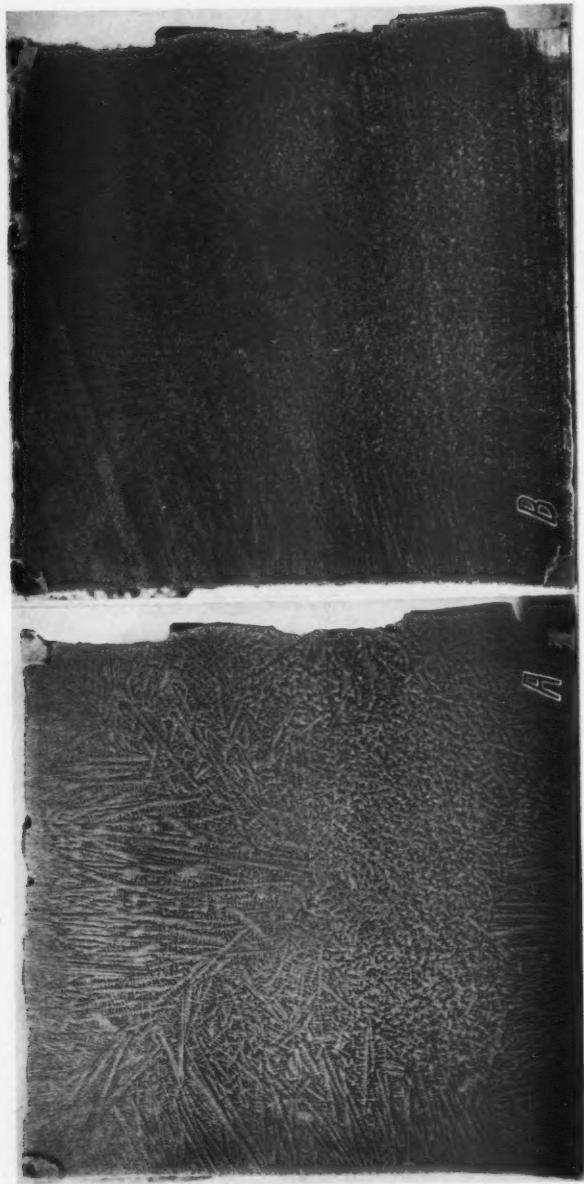


FIG. 1—SECTIONS OF $5\frac{1}{2}$ -INCH SQUARE STEEL CASTINGS ETCHED IN HOT 50% HYDROCHLORIC ACID. *A*: ALLOY STEEL CASTING (0.20 CARBON) WITHOUT VANADIUM. *B*: ALLOY STEEL CASTING (0.20 CARBON) OF SIMILAR COMPOSITION, WITH 0.10 VANADIUM.

The castings portrayed in this illustration were made in the same plant by as nearly as possible identical manufacturing procedure. The second distinction is well illustrated by the photomicrographs of Fig. 2. These also represent the product of the same equipment, personnel and manufacturing practice.

5. While not advocating the use of alloy steel castings with-

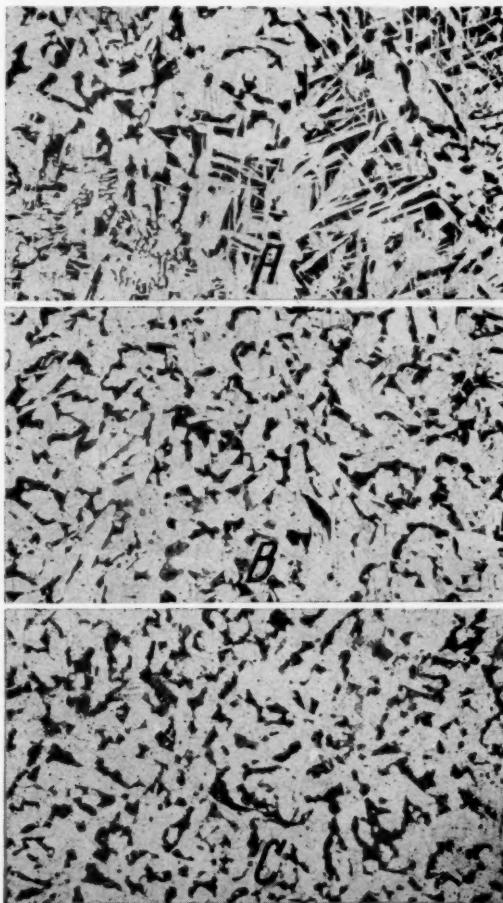


FIG. 2—MICROSTRUCTURE OF 1 1/4-INCH COUPON BARS OF ALLOY STEELS "AS CAST." X35. A: ALLOY STEEL CASTING (0.20 CARBON) WITHOUT VANADIUM. B: ALLOY STEEL CASTING (0.20 CARBON) OF SIMILAR COMPOSITION WITH 0.09 VANADIUM. C: ALLOY STEEL CASTING (0.20 CARBON); ONLY ALLOYING ELEMENT IS 0.14 VANADIUM.

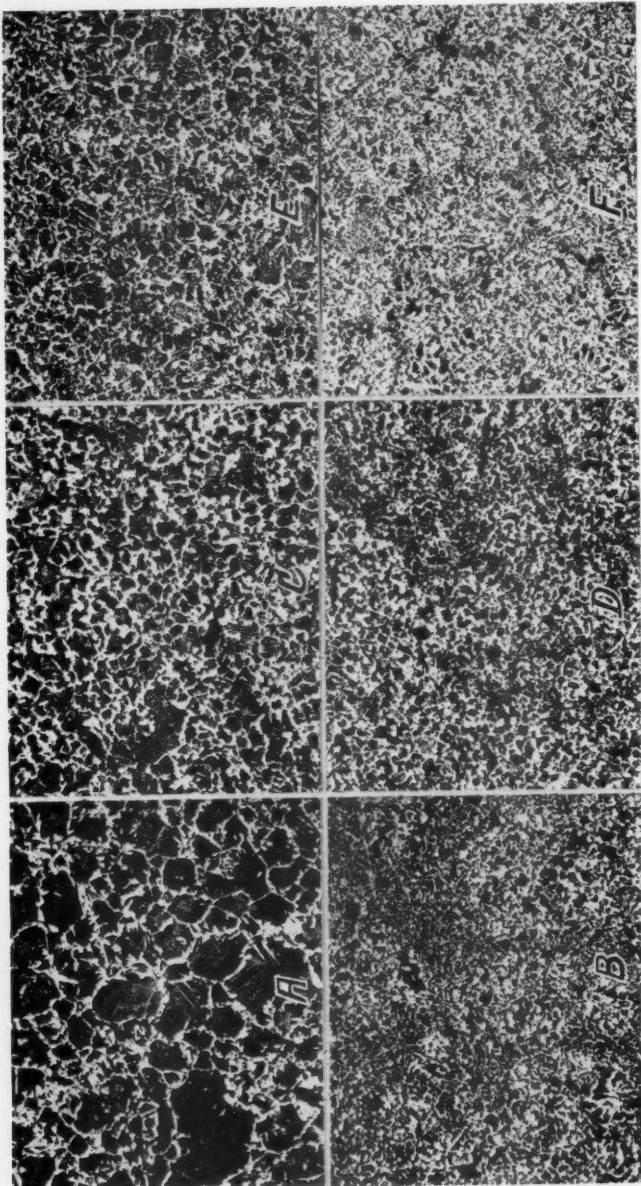


FIG. 3—MICROSTRUCTURE OF 1 1/4-INCH COUPON BARS OF ALLOY STEELS DOUBLE NORMALIZED AND TEMPERED, X100; REDUCED ABOUT ONE-THIRD IN PRINTING. *A*: 1.50 PER CENT MANGANESE STEEL. *B*: 1.50 PER CENT MANGANESE, 0.1 PER CENT VANADIUM STEEL. *C*: 1.00 PER CENT CHROMIUM STEEL. *D*: 1.00 PER CENT CHROMIUM, 0.1 PER CENT VANADIUM STEEL. *E*: 1.50 PER CENT NICKEL STEEL. *F*: 1.50 PER CENT NICKEL, 0.1 PER CENT VANADIUM STEEL.

out heat treatment, it is obvious that for equal hardness and strength, the vanadium steel with its superior distribution of microconstituents possesses greater ability to sustain stress or to deform rather than rupture under suddenly applied overloads. It has frequently been stated that these characteristics are of extreme importance in avoiding the cracking of intricate castings during heating and cooling cycles in the early stages of manufacture.

Response to Simple Heat Treatment

6. The structure of the vanadium-containing steel is also not without influence upon ease of diffusion at heat treatment temperatures. Further, it unquestionably bears a relation to the response of vanadium steels to simple heat treatment. Mechanical properties may be obtained by double normalizing and tempering that are not equaled even by the more drastic and sometimes dangerous quenching and tempering of many other alloy steels. This is of great importance in large complex castings. In fact, suitable adjustment of composition, combined with single normalizing and tempering or single normalizing alone, opens up new possibilities in the use of high-strength cast steels.

7. The macrostructural distinction just noted persists after heat treatment. Microstructurally, equally sharp distinctions develop, causing or accompanied by exceedingly pronounced differences in the mechanical properties. Fig. 3 shows the structure of some alloy steel castings with and without vanadium; both the grain refinement and uniformity of grain size are important.

8. Table 1 is illustrative of the influence of small amounts of vanadium upon the tension and impact values of some normalized alloy steels. The simultaneous increase in both yield point and resistance to impact, and the magnitude of these changes that result from the incorporation of vanadium, without appreciable alteration of tensile strength, elongation or reduction of area, are consistent. The ability to support high static loads and likewise suddenly applied overloads without rupture, disclosed by these comparative tests, has been responsible for numerous and diverse engineering applications.

Carbon-Vanadium Steels

9. The carbon-vanadium steel previously noted has been generally produced within the following composition range:

Table 1
TENSION AND IMPACT TESTS OF CAST ALLOY STEELS DOUBLE NORMALIZED AND TEMPERED

Type*	C	Si	Typical Chemical Composition	Mn	Ni	Cr	Mo	V	Yield Point, lbs. per sq. in.	Tensile Strength, lbs. per sq. in.	Elongation in 2 in., per cent	Reduction of Area, per cent	Impact Value, ft. lbs.
Mn (a)	0.35	0.40	1.40	60,850	102,650	27.5	58.8	25.8
Mn-V (a)	0.35	0.40	1.40	0.10	74,500	100,700	30.5	61.8	57.5
Mn (b)	0.35	0.40	1.40	67,500	108,500	25.9	54.5	19.5
Mn-V (b)	0.35	0.40	1.40	0.10	77,700	103,300	27.5	57.6	52.8
Mn-Mo (a)	0.35	0.40	1.40	0.15	69,300	114,050	21.0	41.6	18.8
Mn-Mo-V (a)	0.35	0.40	1.50	0.15	0.10	...	76,050	118,650	23.0	39.1	30.5
Cr (a)	0.30	0.40	1.80	...	1.00	60,400	94,500	27.5	54.7	38.3
Cr-V (a)	0.30	0.40	0.80	...	1.00	0.10	64,850	94,300	27.5	57.1	59.3
Ni-Cr (a)	0.30	0.40	0.60	1.30	0.50	61,850	92,900	27.0	54.4	48.0
Ni-Cr-V (a)	0.30	0.40	0.60	1.30	0.50	0.10	69,650	94,500	28.5	55.8	71.3
Ni (a)	0.28	0.35	1.00	1.50	53,450	93,250	27.0	56.6	35.5
Ni-V (a)	0.28	0.35	1.00	1.50	0.10	69,900	91,200	28.5	59.4	74.3

* (a) Experimental induction furnace melts. (b) Average of several commercial open-hearth melts heat treated in small furnaces.

Table 2
TENSION AND IMPACT TESTS OF CARBON-VANADIUM CAST STEELS DOUBLE NORMALIZED AND TEMPERED

Melt No.*	C	Si	Chemical Composition	Mn	V	Yield Point, lbs. per sq. in.	Tensile Strength, lbs. per sq. in.	Elongation in 2 in., per cent	Reduction of Area, per cent.	Impact Value, ft. lbs.
1 (b) (c)	0.22	0.40	0.70	0.20	...	46,000	78,000	31.0	58.6	50.2
2 (b)	0.22	0.35	0.92	0.20	...	49,100	80,600	31.0	63.5	63.3
3 (b)	0.25	0.29	0.67	0.19	...	45,600	78,500	24.8	47.4	43.0
4 (b) (c)	0.33	0.33	0.88	0.20	...	52,100	84,800	27.0	55.7	35.0
5 (a)	0.36	0.43	0.67	0.19	...	63,000	90,000	25.0	42.0	35.0
6 (b) (c)	0.37	0.33	0.86	0.20	...	59,900	94,600	24.5	46.0	32.0
7 (b) (c)	0.38	0.32	0.87	0.20	...	64,900	94,100	24.0	53.9	30.0
8 (b) (c)	0.39	0.31	0.58	0.26	...	53,000	82,600	25.0	51.9	22.0
9 (b) (c)	0.40	0.32	0.96	0.18	...	57,400	91,900	26.0	50.6	32.5
10 (b) (c)	0.41	0.31	0.89	0.22	...	56,200	91,400	24.0	41.9	30.8
						53,400	87,100	23.0	47.8	17.5

* (a) Experimental castings. (b) Commercial castings. (c) Single normalized.

Carbon, 0.028 to 0.40 per cent; silicon, 0.25 to 0.40; manganese, 0.65 to 0.90; vanadium, 0.15 minimum. The mechanical properties of castings normalized and tempered are shown in Table 2.

10. It should be mentioned at this point that, with but two exceptions, no test data are included in this paper unless both impact test and tension test results are available. While still imperfectly understood, there has thus far been developed no better test of ability to successfully sustain sudden overloads than the notched-bar impact test. This is of importance to the designing engineer.

11. To metallurgists, this method of examination serves the same purpose and also has the further merit of being a sensitive means for the detection of lack of perfection in steelmaking and molding practice and likewise lack of perfection in heat treatment. From values such as those of Table 2, there have developed a number of specifications of which the tentative specification of the American Society for Testing Materials is representative:

Yield point, lbs. per sq. in.....	55,000 minimum
Tensile strength, lbs. per sq. in.....	85,000 minimum
Elongation in 2 in., per cent.....	22 minimum
Reduction of area, per cent.....	40 minimum

12. The applications for carbon-vanadium steel castings have been principally in locomotive construction and in mining and other heavy machinery. The tonnage has been extremely

Table 3
TENSION AND IMPACT TESTS OF NICKEL-VANADIUM CAST STEELS
DOUBLE NORMALIZED AND TEMPERED

Melt No.*	Chem. Composition— C Si Mn Ni V	Yield Point lbs. per sq. in.	Tensile Strength lbs. per sq. in.	Elongation in 2 in., per cent.	Reduction of Area, per cent.	Izod Value, ft. - lbs.
1 (a) (b)	0.18 0.27 0.84 1.05 0.10	53,850	75,100	32.0	64.7	81.5
2 (c)	0.18 0.38 0.90 1.73 0.08	54,100	89,600	28.5	62.5	72.5
3 (c)	0.20 0.35 0.62 2.11 0.11	93,500	108,000	24.0	62.0	52.0
		71,000	90,500	26.0	64.0	70.0
4 (a) (c)	0.22 0.40 0.70 0.75 0.20	57,000	87,000	30.0	59.4	55.1
5 (a) (c)	0.22 0.40 0.70 1.50 0.20	74,000	94,500	30.0	61.5	59.5
6 (c)	0.22 0.34 0.74 1.55 0.10	56,100	88,700	29.5	58.2	63.0
7 (b)	0.25 0.58 1.01 1.43 0.10	77,000	96,000	28.0	50.0	53.0
8 (c)	0.25 0.39 0.94 1.56 0.11	61,000	92,000	27.0	54.0	60.0
9 (c)	0.27 0.43 1.15 1.71 0.12	100,000	121,000	22.0	52.5	45.0
		82,000	100,000	28.0	62.0	60.0
10 (c)	0.29 0.31 1.02 1.49 0.13	65,000	93,000	29.0	57.8	55.5
11 (c)	0.32 0.34 1.14 1.43 0.12	74,000	103,000	26.0	53.9	56.5
12 (c)	0.32 0.28 1.07 1.58 0.11	70,500	96,500	28.0	59.9	59.5
13 (b)	0.36 0.42 1.00 1.47 0.10	82,000	108,000	25.0	47.0	39.0

*(a) Single normalized. (b) Experimental castings. (c) Commercial castings.

large and the service rendered has been of sufficient merit to warrant continuous use during the twenty-six years that have elapsed since introduction of the composition.

Nickel-Vanadium Steels

13. Of the more complex alloy steels, those containing both nickel and vanadium are among the oldest of industrial importance. Table 3 contains data on the mechanical properties of this type over a rather wide range of carbon content within the limits of constructional utility. Within these ranges of composition, three steels have developed commercial importance and have the following nominal compositions:

Carbon.	Silicon.	Manganese.	Nickel.	Vanadium.
0.20	0.35	0.60	2.00	0.10
0.22	0.35	0.85	1.50	0.10
0.28	0.35	1.00	1.50	0.10

14. A common specification for the third and most important variety, in the normalized and tempered condition, has been as follows:

Yield point, lbs. per sq. in.....	60,000	minimum
Tensile strength, lbs. per sq. in.....	90,000	minimum
Elongation in 2 in., per cent.....	25	minimum
Reduction of area, per cent.....	50	minimum

15. Relatively high values of endurance ratio for cast steels are obtained with this composition, a typical value being 44,000 lbs. per sq. in. endurance limit on coupon bars having a tensile strength of 91,450 lbs. per sq. in., the ratio being 0.482. From this rather unusual combination of high mechanical test values, yield point, impact strength and endurance ratio have developed

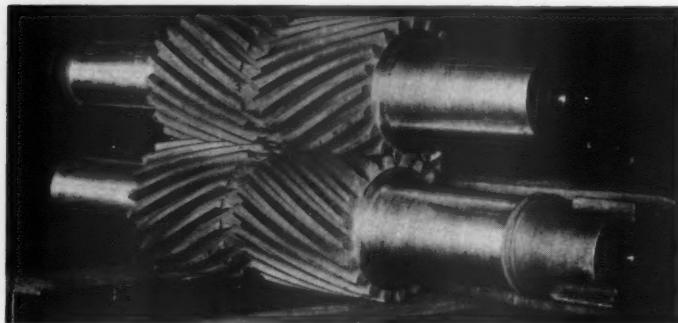


FIG. 4—STEEL MILL PINIONS OF NICKEL-VANADIUM CAST STEEL. WEIGHT OF EACH CASTING, 8000 LBS.

important uses in many industries. Among them may be enumerated engine frames, driving-wheel centers, driving boxes, crossheads, eccentric cranks and crosshead guides of locomotives, gears, pinions, racks and coupling boxes in rolling mills, moving parts of flying shears, hammer rams, ram guides, sow blocks, various types of dies in forging machinery, and a host of other highly stressed parts in many other types of heavy industrial equipment. Fig. 4 shows one of these numerous castings.

16. The transformation temperatures of this steel have been studied, and from the data obtained there are quoted the following values on material conforming in composition to the second type noted above:

— <i>Ac₁</i> —			<i>Ac₃</i>	— <i>Ar₁</i> —		
Beginning.	Maximum.	End.		Beginning.	Maximum.	
1315	1345	1480		1320	1140	1115

Chromium-Vanadium Steel

17. Compositions including both chromium and vanadium have been and are now used to a limited degree but have not spread widely, probably largely the result of lack of early recognition of the air-hardening properties imparted by the chromium.

Table 4
TENSION AND IMPACT TESTS OF MANGANESE-VANADIUM CAST STEELS
DOUBLE NORMALIZED AND TEMPERED

Melt No.*	Chemical Composition				Yield Point, lbs. per sq. in.	Tensile Strength, lbs. per sq. in.	Elongation in 2 in., per cent.	Reduction of Area, per cent.	Isod Value, ft. - lbs.
	C	Si	Mn	V					
1 (e) (a)	0.14	0.30	1.83	0.08	58,600	79,300	33.0	71.0	82.0
2 (b)	0.18	0.30	1.45	0.08	55,400	76,550	35.0	69.3	94.3
3 (b)	0.23	0.48	1.69	0.08	66,050	90,200	32.0	66.0	87.5
4 (b)	0.24	0.42	1.32	0.10	63,000	85,500	32.5	62.0	82.0
5 (b)	0.26	0.41	1.66	0.10	73,500	97,000	29.5	62.0	73.0
6 (b)	0.29	0.42	1.33	0.11	64,350	90,000	31.0	63.5	73.5
7 (b)	0.29	0.41	1.35	0.17	68,000	93,000	29.5	60.0	58.0
8 (b) (d)	0.29	0.42	1.35	0.10	73,000	96,500	28.0	58.0	51.0
9 (b)	0.29	0.46	1.63	0.09	70,800	96,700	30.0	62.3	67.0
10 (b)	0.29	0.38	1.79	0.09	77,800	104,900	27.0	56.5	73.8
11 (b)	0.30	0.44	1.90	0.10	85,500	108,500	27.0	60.0	60.0
12 (b)	0.37	0.48	1.42	0.10	80,500	107,000	25.0	50.0	47.0
13 (b)	0.37	0.36	2.17	0.10	84,750	132,550	14.0	35.6	27.8
14 (e)	0.31	0.51	1.64	0.10	64,180	100,300	28.0	59.0	41.5
15 (e)	0.31	0.48	1.68	0.10	65,500	102,000	26.5	54.1	49.5
16 (e)	0.31	0.49	1.85	0.11	69,000	104,750	27.0	57.5	51.0
17 (e)	0.33	0.36	1.40	0.10	70,500	103,200	25.0	51.9	43.0
18 (e)	0.35	0.40	1.40	0.10	74,500	100,700	30.5	61.8	57.5
19 (e)	0.36	0.43	1.41	0.11	79,750	110,200	26.5	56.0	35.5

*(a) Not tempered. (b) Experimental castings. (c) Commercial castings. (d) Single normalized.

The steels have generally been of a composition approximating the Society of Automotive Engineers' specification 6130, namely, 0.30 per cent carbon, 1.00 chromium and 0.18 vanadium, with 0.70 manganese and the normal silicon content of cast steels. The applications have been in mining and allied equipment, the selection resulting from their excellent resistance to abrasion. That the mechanical properties of the chromium-vanadium steels are good, is clearly indicated in Table 1.



FIG. 5—WATER MAIN TEE OF MANGANESE-VANADIUM CAST STEEL. WEIGHT 13,240 LBS.



FIG. 6—SWAGING MACHINE HOUSING OF MANGANESE-VANADIUM CAST STEEL. WEIGHT, 33,760 LBS.

Manganese-Vanadium Construction Steels

18. The properties of the manganese-vanadium constructional steels, briefly indicated in Table 1, are outlined in somewhat more detail in Table 4, which includes bars from both experimental and commercial melts over the composition range of likely industrial importance, all in the normalized and tempered condition. The influence of vanadium upon the microstructure of pearlitic manganese steels has been noted in Fig. 3 and is so pronounced that the manganese-vanadium steel may be regarded as a totally different product. Sharply contrasted with plain manganese steel, there is observed in the manganese-vanadium type complete absence of that microsegregation so common to pearlitic manganese cast steel and recognized as small martensitic areas under the microscope, which, along with large grain size, contributes to low and variable resistance to impact and poor and irregular machining properties.

19. Complete freedom from these grossly objectionable features of the older product combined with the high mechanical properties has, within a brief period, established for the manganese-vanadium steels a sphere of service closely paralleling that enjoyed by the nickel-vanadium types, namely, as parts of railway equipment and heavy machinery.

20. In Figs. 5, 6, and 7 some of these castings are pictured.

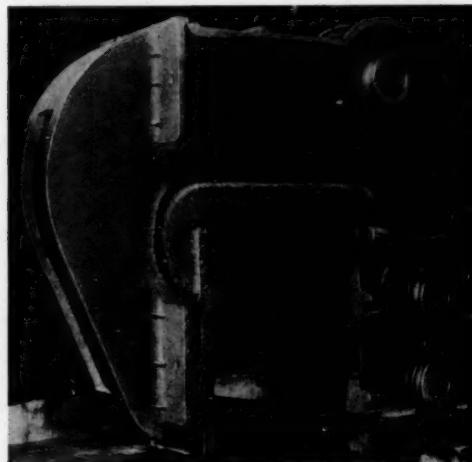


FIG. 7—PUNCH PRESS FRAME OF MANGANESE-VANADIUM CAST STEEL. WEIGHT, 53,070 LBS.

Two composition types have been produced commercially, with essential elements proportioned as follows:

Carbon.	Silicon.	Manganese.	Vanadium.
0.35	0.40	1.40	0.10
0.30	0.40	1.70	0.10

21. In the normalized and tempered condition these steels have been furnished to the same specified minimum tension test values as has the nickel-vanadium cast steel, namely:

Yield point, lbs. per sq. in.....	60,000
Tensile strength, lbs. per sq. in.....	90,000
Elongation in 2 in., per cent.....	25
Reduction of area, per cent.....	50

22. Study of the transformation temperatures has shown, in the case of the higher carbon composition, the following values:

Ac_1		Ac_3	Ar_3	Ar_1	
Beginning.	Maximum.	End.		Beginning.	Maximum.
1350	1385	1480	1340	1215	1190

Some Special Vanadium Steels

23. While the steels just described are the principal ones manufactured in quantity, there are a number of other alloy steel casting compositions containing vanadium used for more or less special applications, all possessing similar qualifications of exceptionally high yield point and impact strength. Tests of a number of these, in the normalized and tempered condition, are listed in Table 5.

Annealed Castings

24. All of the foregoing tests have been restricted to castings and coupons that have been air cooled from above the transformation range and in most instances subsequently tempered. The preponderant tonnage of the castings manufactured in this country are heat treated in this manner. A small percentage of the total production, however, is subjected to other types of heat treatment. Some founders and some purchasers, for example, still cling to the older practice of annealing, while in other instances castings are of such small section and large area, or are so intricate or of such widely varying section, that annealing must be employed to prevent distortion.

25. For guidance in such instances, Table 6 is included, giving the properties in the annealed state of several of the alloy steels previously referred to. It will be observed that the yield ratio and impact strength are somewhat lower than after the more rapid cooling of the normalizing treatment, but that for a

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Table 5
TENSION AND IMPACT TESTS OF VARIOUS VANADIUM CAST STEELS DOUBLE NORMALIZED AND TEMPERED

Melt No.*	Chemical Composition					Yield Point, lbs. per sq. in.	Tensile Strength, lbs. per sq. in.	Elongation in 2 in., per cent.	Reduction of Area, per cent.	Irod Value, It. - lbs.
	C	Si	Min	Ni	Cr	Mo	Cu	V		
1 (b) (a)	0.16	0.26	0.76	0.64	0.08	56,000	77,500
2 (b) (a)	0.19	0.20	0.74	0.57	0.09	65,000	85,200
3 (b)	0.28	0.43	1.44	0.36	0.09	79,000	103,000
4 (b)	0.35	0.37	1.45	0.55	0.10	77,050	108,200
5 (b)	0.43	0.44	1.48	0.80	0.10	85,800	121,200
6 (b)	0.24	0.40	1.32	...	0.61	0.10	70,000	94,000
7 (b)	0.30	0.46	1.51	...	0.83	0.10	82,000	107,500
8 (b)	0.44	0.47	1.48	...	0.61	0.10	76,600	118,250
9 (b)	0.36	0.48	1.17	...	0.33	0.10	90,750	133,450
10 (b)	0.45	0.48	1.53	...	0.16	...	0.10	0.10	83,900	128,900
11 (b)	0.46	0.41	1.37	...	0.66	0.24	...	0.10	142,000	156,600
12 (e)	0.26	0.66	1.15	0.60	0.12	94,000	118,000
13 (a) (e)	0.26	0.40	1.30	0.75	0.20	63,500	92,500
14 (e)	0.31	0.44	1.10	0.90	0.12	81,000	106,500
15 (b)	0.35	0.43	1.43	0.62	0.10	94,500	113,000
16 (b)	0.35	0.91	1.56	0.10	85,000	100,000
									112,000	126,000
									111,050	111,050
									25.0	25.0

* (a) Single normalized. (b) Experimental castings. (c) Commercial castings.

Table 6
TENSION AND IMPACT TESTS OF VARIOUS VANADIUM CAST STEELS ANNEALED

Melt No.*	Chemical Composition					Yield Point, lbs. per sq. in.	Tensile Strength, lbs. per sq. in.	Elongation in 2 in., per cent.	Reduction of Area, per cent.	Irod Value, It. - lbs.
	C	Si	Mn	Ni	V					
1 (b)	0.14	0.30	1.83	...	0.08	50,700	68,000	37.0	73.7	65.3
2 (b) (d)	0.32	0.49	1.57	...	0.11	64,400	93,200	29.5	56.9	53.0
3 (a)	0.34	0.30	0.82	...	0.09	47,500	78,350	30.5	54.3	48.8
4 (b)	0.36	0.43	1.41	...	0.11	60,000	96,300	26.8	50.0	28.0
5 (b)	0.28	0.39	1.34	1.72	0.11	71,500	103,050	23.0	40.4	43.3
6 (b) (c)	0.29	0.32	0.98	1.68	0.12	58,000	87,600	20.5	54.8	57.0
7 (b)	0.30	...	0.95	1.50	0.14	61,750	95,500	24.0	44.5	28.0
8 (b)	0.39	0.30	1.05	1.82	0.14	67,050	105,400	26.0	46.3	27.8

* (a) Experimental castings. (b) Commercial castings. (c) Previously normalized and annealed.

given hardness or tensile strength these values are still well above those of most cast steels not containing vanadium.

Quenched and Tempered Castings

26. A small percentage of the product of the steel foundries is also used in the quenched and tempered condition. While not to be recommended for castings other than those of simple form and uniform section, this method of heat treatment may result, if intelligently prescribed and executed, in castings having rather high strength properties with relatively good impact value and, hence, be suited to service requiring hardness, extreme wear resistance and ability to sustain suddenly applied loads.

27. Table 7 contains the results of a number of tests of coupons heat treated in this manner. The impact test values are high in relation to the tensile strengths, while in the lower ranges of hardness it will be noted that yield ratios higher than those of the normalized steels have been developed. In the higher hardness ranges, namely 150,000 to 200,000 lbs. per sq. in., numerous castings have been made that have been exceptionally serviceable in the wearing parts of crushing machinery, equipment for handling abrasive substances, and other similar uses.

Molybdenum-Vanadium Steel for Centrifugally Cast Guns

28. A discussion of quenched and tempered castings would be incomplete without reference to the centrifugally cast, heat-treated and cold-worked guns developed at the Watertown Arsenal. Following trial of a number of alloy steel compositions, the molybdenum-vanadium type represented by melt No. 3 of Table 8 was selected, presumably the result of its consistently fine grain size (both primary and secondary), its excellent mechanical properties and its ability to sustain a maximum of cold deformation without rupture or loss of ductility. It is unfortunate that the use of a special form of test prevents comparison of the impact data of Table 8 with those in the remaining tables of this paper. On the basis of the above comparison of steels, however, it must be assumed that these values are most satisfactory for this application.

Behavior of Steels as Heavy Castings

29. Other than these tests of a molybdenum-vanadium steel for ordnance, which were taken from midway positions in sections

Table 7
TENSION AND IMPACT TESTS OF VARIOUS VANADIUM CAST STEELS NORMALIZED, QUENCHED AND TEMPERED

Melt No.*	C	Si	Mn	Ni	Cr	Mo	V	Chemical Composition	Yield Point, lbs. per sq. in.	Tensile Strength, lbs. per sq. in.	Elongation in 2 in., per cent	Reduction of Area, per cent	Iso'd Value, ft.-lbs.
1 (b)	0.29	0.46	1.72	0.10	100,000	173,700	11.5	41.5	13.5	
								135,150	145,100	14.5	44.3	21.0	
								101,800	112,800	22.5	56.2	43.0	
2 (a)	0.29	0.38	1.70	0.09	111,300	125,400	21.0	61.2	43.0	
3 (b)	0.36	0.43	1.41	0.11	170,000	183,600	11.0	35.4	11.0	
								144,100	155,000	19.0	38.8	19.0	
								103,550	114,300	20.5	60.6	42.5	
4 (b)	0.26	0.30	0.97	1.54	0.11	91,150	102,400	26.0	58.1	49.5	
								116,700	140,000	15.5	46.0	23.5	
								107,500	121,750	18.0	43.1	34.5	
5 (b)	0.29	0.32	0.98	1.68	0.12	94,900	106,800	24.0	56.5	37.5	
6 (b)	0.30	0.27	1.02	1.62	0.12	100,150	131,650	18.0	46.4	34.0	
								79,650	94,300	26.5	63.0	63.8	
								131,500	153,000	13.0	36.3	18.5	
7 (b)	0.31	0.44	1.10	0.90	0.12	101,200	113,400	21.0	53.6	43.8	
8 (a)	0.33	0.43	1.04	1.57	0.10	83,000	96,500	28.5	63.3	67.0	
9 (a)	0.33	0.41	1.41	0.80	0.27	0.10	123,000	146,350	14.0	42.0	23.0
10 (a)	0.38	0.44	1.35	0.78	0.25	0.10	163,000	173,000	12.0	29.0	18.0
								158,000	150,000	16.0	41.0	25.0	

* (a) Experimental castings. (b) Commercial castings.

Table 8
TENSION AND IMPACT TESTS OF MOLYBDENUM-VANADIUM CENTRIFUGALLY CAST STEEL NORMALIZED, ANNEALED, QUENCHED AND TEMPERED

Melt No.	Chemical Composition				Location	Proportional Limit, lbs. per sq. in.	Tensile Strength, lbs. per sq. in.	Elongation in 2 in., per cent	Reduction of Area, per cent	Charpy Impact (a), ft.-lbs.
1	0.51	0.31	0.05	Breech	75,000	99,000	26.7	58.3	40.0
2	0.42	0.30	0.05	Musale	73,000	101,000	25.5	57.9	43.0
					Breech	73,000	98,000	26.5	57.6	40.0
					Musale	76,000	95,000	28.9	60.7	41.0
3	0.35	0.29	0.10	Breech	68,000	91,000	51.9	39.0
					Musale	77,000	96,000	60.6	36.0
4	(b)†	Breech	72,000	101,000	21.2	52.7	33.9
					Musale	70,000	97,000	24.9	57.4	36.9
5	(b)†	Breech	72,000	97,000	22.4	49.9	35.7
					Musale	73,000	99,000	22.3	47.6	37.8

* (a) Special tension specimen in Charpy Machine.

† (b) Analyses supposedly similar to that of Melt No. 3.

ranging from about 2 to about 5 inches in thickness, all of the foregoing data have been secured on coupons 1 to 1½ inches thick. While many represent the steel of separately poured test blocks, others have been attached to commercial castings of different weights, some light, others heavy.

30. It is important, however, to know how steels selected for specific uses will behave when applied to the manufacture of heavy castings which heat and cool relatively slowly both after solidification and during heat treatment, this knowledge to include not only the properties of attached test bars but also those of the surface and of the interior of heavy masses. In the latter, the properties are influenced, of course, not only by the behavior of steel of given composition when subjected to a variety of heating and cooling rates but also by the soundness of the mass as controlled by molding and casting practice.

31. In the case of sound metal, it is enlightening to examine the properties of large masses of vanadium steels as included in Table 9. It is evident from these data that the vanadium steels previously described develop, in test bars attached to very heavy castings, tension and impact properties readily meeting exacting specification requirements prepared largely from tests of smaller units, and that even bars cut from the body of very heavy sections exhibit properties differing but slightly from those of standard test coupons.

Properties at Sub-Normal Temperatures

32. The mechanical properties quoted thus far have all been determined at room temperature; most castings are used under conditions that make this information of primary importance. For a few uses, however, the properties at sub-normal temperatures are of importance. The only published data of this sort at low temperatures, of vanadium steel castings, are due to Moffatt, who determined the impact strength of cast carbon and a few alloy steels at temperatures down to those prevalent in the more northern civilized countries. He found that within the range of +68 to —36 degs. Fahr., a nickel-vanadium steel of the type here described gave the maximum impact strength at all temperatures of the range and a minimum difference of 10.6 per cent between the values which he was able to obtain at the extreme temperatures.

Table 9
TENSION AND IMPACT TESTS OF LARGE SECTIONS OF VANADIUM STEEL CASTINGS DOUBLE NORMALIZED AND TEMPERED

Melt No.	Chemical Composition —				Section Size and Location *	Yield Point, lbs. per sq. in.	Tensile Strength, lbs. per sq. in.	Elongation in 2 in., per cent.	Reduction of Area, per cent.	Fracture	Izod Value, ft.-lbs.	
1	0.32	0.28	0.69	0.17	5 $\frac{1}{4}$ in. sq. corner.	53,200	80,100	29.5	58.8	Cup	20.5	
					5 $\frac{1}{4}$ in. middle of side.	54,300	80,080	25.0	45.2	$\frac{1}{2}$ Cup	21.1	
2	0.40	0.33	0.75	0.17	5 $\frac{1}{4}$ in. sq. corner.	56,450	86,850	27.5	55.7	18.0	
					5 $\frac{1}{4}$ in. middle of side.	61,980	88,500	27.0	55.2	18.0	
3	0.30	0.38	1.65	0.11	Coupon (c)	61,750	90,150	24.5	48.4	17.0	
	0.36	0.38	1.40	0.12	(d)	60,700	88,500	23.0	40.1	16.0	
4	0.22	0.34	0.74	1.55	0.10	5 $\frac{1}{4}$ in. sq. corner.	68,500	103,260	27.0	61.8	48.0
					5 $\frac{1}{4}$ in. center (b)	54,300	86,700	28.5	52.5	29.5	
5	0.23	0.39	0.91	1.56	0.11	5 $\frac{1}{4}$ in. sq. corner.	60,450	85,150	29.5	54.1	31.0
					5 $\frac{1}{4}$ in. middle of side (a)	59,600	85,450	30.5	54.7	27.0	
6	0.23	0.39	0.91	1.56	0.11	5 $\frac{1}{4}$ in. sq. corner.	57,850	79,100	30.0	63.0	1 Cup	62.0
					5 $\frac{1}{4}$ in. middle of side (a)	52,250	78,350	28.5	57.0	$\frac{1}{2}$ Cup	60.5	
7	0.31	0.35	1.00	1.65	0.12	Coupon (e)	58,100	78,750	29.0	54.7	$\frac{1}{2}$ Cup	68.5

* (a) Middle of vertical side; not cone or drag. (b) Center of test bar 1 in. from center of block. (c) Test block attached to middle of casting 14 in. thick, weighing 10,000 lbs. (d) From Vee Guides of 8000-lb. hammer ram 34 in. thick. (e) Test block attached to hub casting weighing 28,000 lbs.

Properties at Elevated Temperatures

33. At temperatures above atmospheric, no conclusive data from either short or long-time tests are available on vanadium cast steels. However, with tests upon wrought materials as a basis, work has been done with several low-alloy steel types such as chrome-vanadium, tungsten-chrome-vanadium and manganese-molybdenum-vanadium. These uses, with the exception of the application of several compositions of tube-mill piercing points, are not yet old enough nor diverse enough to permit definite statements of utility at this time; but from the results obtained thus far and the test data upon wrought materials, much is with reason expected of them.

Cast Tool Steels

34. Some criticism may be offered of the omission of cast steel dies and other tools from this discussion. The use of vanadium in such compositions, however, is so widespread that detailed inclusion of them here appears unwarranted. It seems sufficient to indicate that the same advantages that are conferred upon the low-alloy constructional steels are of as great and sometimes greater importance in both the low-alloy and high-alloy higher carbon steels used in the hardened state.

35. Improved distribution of phases, both macroscopically and microscopically, uniformity of successive melts and increased toughness and useful strength, are characteristic. Improved diffusion during heat treatment, and control of depth of hardening, also are observed. Thus, in addition to its indispensable presence in cast high-speed steel tools, vanadium is employed in most of the high-carbon high-chromium die steels, in high-carbon steel rolls along with chromium, and in many medium-carbon cast die steels such as the nickel-chromium-molybdenum-vanadium type for hot working.

Applications for Corrosion and Heat Resistance

36. Applications to highly alloyed castings of the corrosion and heat-resistant class have been few, but it may yet be used in these steels for its influence upon grain size and its restraint of deterioration upon long exposure to elevated temperatures. The next few years may also witness commercial use of vanadium-bearing cast steels surface hardened by nitriding. Some work has been done in this direction indicating that the absence of aluminum

in these vanadium steels avoids the disadvantages of previous nitriding compositions, while retaining the good qualities inherent in nitrided articles. At present, however, the development stage has not been passed.

ACKNOWLEDGMENTS

37. The mechanical test values recorded in this paper have been accumulated from many sources. In addition to data obtained in the laboratories of the Vanadium Corporation of America, many have been contributed through private communications by W. C. Hamilton of American Steel Foundries, Richard Armstrong of Union Steel Casting Co., R. B. Tillman of Industrial Steel Castings Co., and Fred Grotts of Continental Roll & Steel Foundry Co. As previously noted, selections of test data have been limited by the desire for the inclusion only of results embodying details of composition and resistance to impact as well as tension test values, but the Bibliography appended hereto has nevertheless been able to provide much that has been of real assistance.

(For discussion of this paper see page 256)

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Chromium in Steel Castings

By J. H. CRITCHETT,* NEW YORK

1. Before taking up the specific role of chromium in the steel casting industry, it may be well to say a few words about the increasing importance of alloy steels in general in that industry. With the feeling that there may not be a general appreciation of the steadily increasing part that alloy steel castings are playing, figures have been gathered from 1921 to date showing the climb of the industry out of the post-war depression. In Fig. 1, one curve shows the total steel casting production; the second, the alloy steel casting production, and the third, the percentage relationship which alloy steel castings bear to total steel castings.

2. It will be noted that the total steel casting production very quickly reached a normal value and held fairly steadily for several years until the present difficulties came upon us. On the other hand, the percentage of alloy steel castings climbed steadily, making a new high each year. There was a slight drop in percentage in 1929-1930 but the progress was then resumed even in the dull years following. From a figure of only 4 per cent, the proportion of alloy steel castings has increased to over 17 per cent and gives every indication of going further.

3. There are many reasons for the rapidly increasing importance of alloy steel castings, based on the ever-increasing tempo of production and the consequent demand for machinery and structures capable of withstanding severe duty. Not the least important reason, however, is the better understanding of the problems of heat treating on the part of the foundry personnel and the more general equipment of foundries to do this kind of work. Once heat treatment became generally applied to all plain carbon steel

* Vice-President, Union Carbide and Carbon Research Laboratories.

NOTE: This paper was presented and discussed before the steel alloys session at the 1933 Convention of the American Foundrymen's Association. The discussion of this paper will be found beginning on page 256 where the discussion of the four papers presented at this session is reproduced.

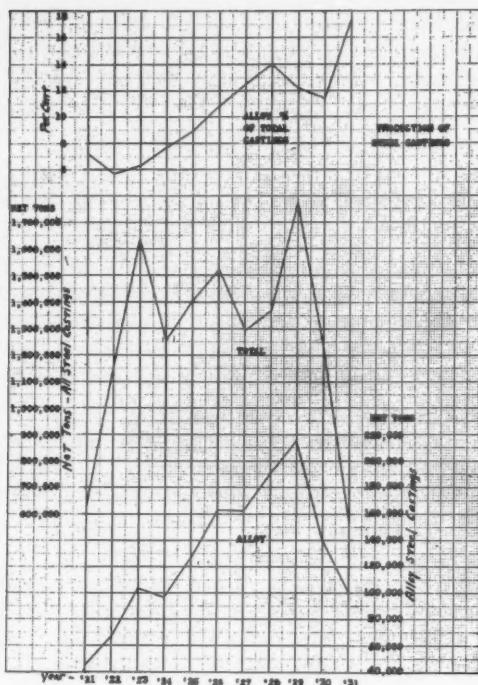


FIG. 1.—PRODUCTION OF STEEL CASTINGS, 1921 TO 1931.

castings, analyses were found for alloy steel castings which fitted readily into steel foundry practice, and whose alloy cost added only slightly to the cost of the liquid steel and even less relatively to the cost of the final casting. When the superior properties that can be obtained through this means are considered, the increased cost of the alloy steel casting is returned many times to the purchaser through the better service which he obtains from it.

4. The steel founder has a wide range of choice in the particular alloying element or elements he may use. Each element recommends itself for consideration, and the advantage of special properties must be balanced against the cost of production.

5. The purpose of this paper is to present the case for chromium castings. It will be remembered that chromium is used in conjunction with numerous other alloying elements which supplement the advantages of chromium taken singly. Among these

are nickel, molybdenum, vanadium, manganese, silicon and others, one or more of which may be used to augment the mechanical or develop special chemical qualities in the steel. The complex steels do not come within the province of this article, which deals only with plain chromium steels.

6. In considering plain chromium steels there is a tendency to lose sight of the excellent mechanical properties that may be obtained with this element alone, in the light of the more spectacular behavior of the somewhat higher chromium steels, with regard to corrosion. This is a mistake, since the plain chromium steels of low chrome content have physical properties well adapted to many structural needs.

7. Since the chrome steels grade from pure structural to corrosion resistant, the present paper will be divided into three parts. The first deals with castings containing 1 to 3 per cent chromium used for structural parts. The second covers a chromium content of from 4 to 7 per cent, used where both structural and certain corrosion requirements must be met. In the third part some cursory mention will be made of the special high-chromium corrosion-resistant steels.

I—STEELS OF 1 TO 3% CHROMIUM FOR STRUCTURAL PARTS

8. In Table 1 selected data are set down to illustrate physical properties that can be obtained from plain chromium steel castings with ordinary heat treatments. The carbon, silicon and manganese of the five steels illustrated are sufficiently uniform so that from the point of view of analysis chromium is the variable. The first is a plain carbon steel treated for the best combination of strength and ductility.

9. The second illustrates the fact that by dropping the carbon slightly and adding a small amount of chromium, both strength and ductility are raised. The point is important that the carbon is lowered and deserves some elaboration. Often in dealing with alloy steels the metallurgist works from the carbon steel end and attempts to modify the carbon steel by alloy additions. In reality, he will get better results by considering the alloy combination first and adjusting the carbon to give the desired results. By this route, in general, maximum ductility will be obtained for a given strength, and by the simplest heat treatments. This fact is still further illustrated by the third steel in this group which, if it had slightly lower carbon and higher chromium, would have had equally high strength and better ductility.

Table 1
PROPERTIES OF CAST STEELS OF LOW CHROMIUM CONTENT

C	Mn	Si	C.R.	Physical Properties				Treatment* (temp. in degs. Fahr.)
				Ultimate Strength lbs. per sq. in.	Yield Point lbs. per sq. in.	Elongation in 2 in. Per Cent	Reduction of Area, Per Cent	
0.32	0.82	0.33	...	77,000	43,000	32	57	N.
0.26	0.90	0.68	0.65	84,000	51,000	33	63	1650 A.C.
0.38	0.74	0.50	0.81	97,000	55,000	19	30	1650 A.C.
0.36	0.41	0.23	3.00	118,000	94,000	23	51	1650 A.C.
0.36	0.78	0.78	...	151,000	127,000	14	29	1650 W.Q.
0.36	0.78	0.78	2.97	142,000	122,000	15	34	1600 O.Q.
0.36	0.78	0.78	2.97	138,000	120,000	16	37	1500 O.Q.
			...					1150 A.C.

* N. = normalized; D. = drawn; A.C. = air cooled; W.Q. = water quenched; O.Q. = oil quenched.

Table 2
PROPERTIES OF LOW-CHROMIUM STEELS OF CROMANSIL ANALYSIS

C	Mn	Si	Cr	V	Physical Properties				Treatment* (temp. in degs. Fahr.)
					Ultimate Strength lbs. per sq. in.	Yield Point lbs. per sq. in.	Elongation in 2 in. Per Cent	Reduction of Area Per Cent	
0.37	1.49	0.46	0.52	...	117,000	73,000	20	48	1650 A.C.
0.39	1.38	0.57	0.63	...	124,000	75,000	21	49	1650 A.C.
0.42	1.45	0.80	0.38	0.16	116,000	74,000	21	41	1650 O.Q.
0.42	1.45	0.80	0.38	0.16	116,000	74,000	19	43	1650 A.C.
0.42	1.45	0.80	0.38	0.16	105,000	73,000	26	49	1650 O.Q.
0.42	1.45	0.80	0.38	0.16	108,000	62,000	20	47	1650 A.C.
									1450 A.C.

* A.C. = air cooled; Anl. = annealed; O.Q. = oil quenched.

10. The last two analyses in Table 1 are characterized by somewhat unusually high chromium content. The physical properties are very good, the ductility being high for steel of such high yield point and ultimate strength. Steels such as these are eminently suited to some structural uses, particularly where resistance to abrasion is needed.

11. Table 2 contains data on another type of low-chromium steel of the Cromansil variety. In this analysis the manganese and silicon are about 0.5 and 0.25 per cent, respectively, over that common for plain carbon steel castings, and the chromium is held around 0.5 per cent. The steels are easily made and call for only a very moderate alloy cost. They run freely and behave well in the foundry, and they can be heat treated by ordinary air cooling methods. The ductility figures are exceptionally good for such high strength, being better than would be expected of plain carbon steels liquid quenched.

12. The last four sets of figures in this table are for a Cromansil analysis to which a small vanadium addition has been made. The main effect of this addition is to obtain grain refinement, since its presence does not materially affect the physical properties. The figures are given for this analysis heat treated in four different ways. It will be noticed that while oil quenching gives slightly higher ductility values than air cooling, the improvement is hardly sufficient to warrant the additional cost and bother for the oil-quenching method, except when rigid specifications must be met. The main object of the inclusion of these figures is to emphasize the fact that by proper alloy combinations, steels can be produced whose properties are developed by simple, inexpensive methods of heat treatment.

13. Only a few technical data have been given to illustrate the wide range of properties that can be obtained with chromium alone. Unfortunately, all the properties of a steel casting cannot be set down in a simple tabulation of figures; hence, such a table does not tell the whole story.

14. The chrome steels have a combination of toughness and hardness which makes them very suitable for wear-resistant parts. In particular, the 3 per cent chromium grade is extremely well suited to such parts as rock-crushing equipment and dipper and dredge teeth. Chromium imparts a unique depth-hardening quality which simplifies heat treatment and insures good uniform properties throughout massive sections and throughout castings of greatly varying section. Considering the fact that chromium has

Table 3
PROPERTIES OF STEELS WITH HIGHER CONTENT OF CHROMIUM†

C	Mn	Si	Cr	W	Mo	Ni	Physical Properties			Treatment*	(Temp. in degs. Fahr.)
							Ultimate Strength lbs. per sq. in.	Yield Point lbs. per sq. in.	Elongation in 2 in. Per Cent		
0.25	0.53	0.42	5.87	0.80	125,000	100,000	18	45	N. 1750, D. 1250
0.22	0.71	0.60	4.02	0.46	121,000	91,000	19	51	N. from 1650
0.12	0.30	0.45	13.56	97,000	72,000	20	42	1650 A.C.
0.30	0.63	0.78	18.56	104,000	73,000	13	13	
0.32	0.73	0.59	25.46	85,000	55,000	7	7	
0.42	0.76	0.45	24.30	0.22	106,000	71,000	12	12	As Cast
0.44	0.71	0.51	24.80	0.05	84,000	52,000	7	7	As Cast

† "Properties of Low-Carbon Medium-Chromium Steels of the Air-Hardening Type," by E. C. Wright and P. F. Mumma, metallurgists, National Tube Co., American Institute of Mining and Metallurgical Engineers, February 1933.

* N = normalized; D = drawn; A.C. = air cooled.

the lowest cost of any of the alloying elements not regularly used in steels, it is evident that through its use the metallurgist can often obtain the maximum in properties at the minimum expense.

II—4 TO 7% CHROMIUM FOR STRUCTURAL AND CORROSION REQUIREMENTS

15. This section deals with the steels of 4 to 7 per cent chromium content. The physical properties of this class at room temperature are sufficiently well shown in Table 3. The properties which are creating a new market for the 4 to 7 per cent chromium steels are: Strength at moderate temperature, noticeable increase in corrosion resistance, and resistance to scaling at moderate tem-

Table 4
RESISTANCE OF 0 TO 7 PER CENT CHROMIUM STEELS TO CORROSION PENETRATION IN 30-DAY TEST

Test Solutions.	Chromium Content of Steels, per cent				
	0.00	1.06	2.77	4.64	7.14
	Inches Penetration in 30 Days				
Hot Aerated City Water	0.20	0.030	0.010	0.009	0.011
Aerated Dilute HCl	0.025	0.022	0.008	0.005	0.004
Aerated NaCl Solution	0.012	0.009	0.007	0.007	0.008
NaCl Solution Aerated with H ₂ S	0.023	0.010	0.005	0.004	0.003
Aerated Dilute H ₂ SO ₄	0.024	0.020	0.013	0.010	0.007

perature. True creep strength figures are not yet available, but short-time tests indicate a safe load in excess of 20,000 lbs. per sq. in. at 800 degs. Fahr., and experience has shown very satisfactory performance at even higher temperatures. The corrosion resistance of this class of steels is indicated by the test results of Table 4, showing inches of penetration in a 30-day test.

16. While these steels cannot be compared with the higher chromium true stainless steels for corrosion resistance, the tabulation shows a worth-while increase in corrosion resistance with 3 per cent chromium or better in all tests, and a very good increase in resistance against dilute hydrochloric acid, hydrogen sulphide aerated salt, and dilute sulphuric acid solutions.

17. The oxidation resistance of this type of steel has been sufficiently confirmed in practice to give assurance that it is com-

Table 5
RELATIVE OXIDATION RESISTANCE

Sample.	Per Cent. Loss in Weight.
Ship Plate	34
Cast Iron	36
5.47 Cr, 0.11 C.....	20
6.0 Cr, 0.13 C, 0.90 Ti.....	11
6.8 Cr, 0.06 C, 0.69 Ti.....	2

mercially valuable for such service. The figures of Table 5 were obtained on samples which were all held in an open muffle furnace at 1500 degs. Fahr. for four weeks and the percentage loss in weight then determined.

18. It will be noticed that 6 per cent chromium steel could be expected to have materially longer life than ordinary steel at this temperature. The addition of a fairly large amount of titanium still further increases the life of the plain chromium steel, and if the carbon is held very low and a titanium addition made, the life becomes remarkably long. The development of the addition of titanium to the plain chromium steels to the extent shown in the example is by the Union Carbide and Carbon Research Laboratories, Inc.

19. The 4 to 6 per cent chromium steels without titanium addition may have a very low elongation and reduction of area, depending on the way in which the castings have cooled. This is due to the extreme air-hardening properties imparted by the chromium. Titanium has the effect of almost entirely eliminating the air-hardening property so that the steels are relatively ductile as cast. Simple reheating to about 1650 degs. Fahr. with slow cooling still further increases the ductility and toughness to values comparable with low-carbon ordinary steel castings.

20. The sluggishness and poor foundry qualities that would naturally be expected from a steel with so much chromium and so little carbon can be overcome by increasing the silicon above that usually employed; the added silicon apparently has no detrimental effect on the stability of this steel under corrosive conditions. Other elements are known to bring about the same effect as titanium, and these elements have a similar beneficial effect on other chromium steels than those containing 4 to 7 per cent of chromium. Unquestionably, further commercial experience with this modified analysis will greatly expand the field of use of the plain chromium steels.

III—SPECIAL HIGH-CHROMIUM CORROSION-RESISTANT STEELS

21. No discussion of the plain chromium steels would be complete without reference to the high-chromium corrosion and heat-resisting steels. However, so much has been written on this subject that here we will attempt only to summarize briefly the voluminous articles that have appeared elsewhere. Strauss,* only a year ago for this same society, gave such a complete story about these steels that little can be added to it. For the purpose of this present discussion the plain chromium corrosion and heat-resisting steels will be divided into three groups, those containing from 12 to 16 per cent, 17 to 20 per cent, and upward of 20 per cent chromium. A typical set of figures indicating the physical properties of each of these grades is included in Table 3.

22. The effect of chromium in increasing the corrosion resistance of steel is sufficiently marked at about 4 per cent chromium so that there is an important class of steel made for this service containing 4 to 7 per cent chromium which has been previously referred to. At about 11 per cent chromium content there is very marked increased resistance, so much as to create a very important class of commercial corrosion-resisting steel containing 12 to 14 per cent of that element. Castings of this steel are very resistant to corrosive action by fruit juices, hot lubricating oil, vinegar and vegetables, also basic materials such as lime and caustic soda. Valves cast from them have given excellent service in high-pressure steam lines and hot oil transfer lines.

23. By increasing the chromium content to about 18 per cent, corrosion resistance is still further improved. Castings in this class have given good service in the form of pump valves, valve trim, superheated steam lines, resistance to sea water, hot oils, foodstuffs and alkaline solutions, both hot and cold. Steels of still higher chromium content are used more particularly to resist extremely high temperature conditions, as in the handling of hot gases.

24. Resistance to oxidation at elevated temperatures improves in a somewhat lineal manner up to about 11 per cent chromium and is very markedly increased by still larger amounts. The 12 to 14 per cent chromium castings can be used successfully for long periods of time in heat interchangers, furnace parts and similar equipment at temperatures up to about 1400 degs. Fahr. Those containing about 18 per cent chromium function very well at tem-

* Jerome Strauss, "Castings of Corrosion Resistant Steels," A.F.A.—A.S.T.M. Symposium on Steel Castings, *Proceedings A.S.T.M.*, vol. 32, pt. 2, 1932.

peratures up to 1650 degs. Fahr., but where resistance to scaling is required, steels containing upward of 20 per cent and in some cases up to double this amount of chromium are much to be preferred. Their resistance to scaling is very good up to 1800 degs. Fahr. and often to as high as 2000 degs. Fahr.

25. Notable applications for this class of steel may be found in supporting disks in steel heat-treating furnaces, waste-heat re-eupernators, superheaters containing corrosive gases at high temperatures, rolling mill guides, and numerous other uses. They have been very successfully used as retorts to regenerate bone char in the sugar industry; in the form of pug-mill blades, tubes, hangers and saddles in the cement industry; for the handling of mixed acid solutions in the manufacture of gunpowder, and are without a peer in their performance in applications requiring resistance to acid mine water.

26. One of the difficulties heretofore encountered in the casting and use of steels containing over 20 per cent chromium is the tendency toward large columnar grains in the castings, and toward grain growth when held at high temperature for very long periods. The first difficulty has been entirely overcome and the second very materially reduced through the addition of appropriate amounts of nitrogen.

27. Nitrogen is added through the medium of a special high-nitrogen ferrochrome, and must be present in the steel to the extent of one part for every hundred of chromium. Its use increases the strength and toughness very materially and the hardness slightly, but without detrimental effect on machinability.

28. The particular value of the nitrogen addition, however, is in very greatly decreasing grain size, with consequent increase

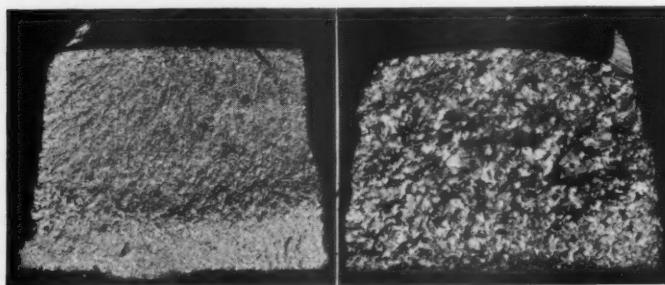


FIG. 2—GRAIN REFINING EFFECT OF NITROGEN ADDITION TO 25 PER CENT CHROMIUM STEEL CASTINGS.

in toughness. The striking effect is well illustrated in photographs (Fig. 2) which show steels from heats as identical as possible in all respects except one containing fine crystals, this having received the nitrogen treatment.

29. In Table 3 are figures comparing the physical properties of two 25 per cent chromium castings, one containing normal and the other the special amount of nitrogen. The marked increases in ultimate strength and yield point are easily apparent, but particular attention is called to the figures for elongation and reduction of area which, in the nitrogen-treated steel, are approximately double those of the steel with ordinary nitrogen content. These figures are for the "as cast" condition and could be somewhat improved by proper heat treatment.

30. Previous publications of the American Foundrymen's Association and the files of Union Carbide and Carbon Research Laboratories, Inc., have been drawn on for data without special acknowledgment. Grateful acknowledgment is given the American Steel Foundries, Atlantic Steel Castings Co., Empire Steel Castings Co., Inc., and the Duraloy Company for their kindness in allowing the use of their data for appropriate parts of this paper.

(For discussion of this paper see page 256)

Symposium—Alloys in Cast Steel

At the session on alloys in steel castings, held June 21, 1933, the following four papers were presented:

Molybdenum in Cast Steel, by H. W. Gillett and J. L. Gregg, Battelle Memorial Institute, Columbus, O.

Properties and Uses of Some Cast Nickel Alloys, by Albert G. Zima, International Nickel Co., New York.

Vanadium in Steel Castings, by Jerome Strauss, Vanadium Corp. of America, Bridgeville, Pa.

Chromium in Steel Castings, by J. H. Critchett, Union Carbide and Carbon Research Laboratories, New York.

The meeting was called to order by Chairman R. A. Bull and after presentation of the papers by the authors, the following discussion took place:

CHAIRMAN R. A. BULL¹: The four papers presented cover a large number of grades of alloy cast steel. There probably are 50 different varieties of that material that are made by intentional introduction of one or more of these four alloys. I think most steel foundrymen desire further information concerning these and other alloys that have been used in making special cast steels.

Five years ago, there was a greater tendency in foundries to confine their work to a few special types of alloy steels than there is today. The tendency to produce, in one plant, a great many compositions may become less as we learn more about the results from varied heat treatments. We can get a large variety of physical properties from each of many alloy steel compositions, depending on the heat treatment used.

F. G. FRISBIE²: (*Presented in written form and read by Chairman Bull*) I would like to offer a few notes in conjunction with the paper, "Molybdenum in Cast Steel". I have had experience in the manufacture of nickel-molybdenum, chromium-molybdenum, manganese-molybdenum, chromium-manganese-molybdenum and a little in vanadium-molybdenum castings. Any of those readily can be treated to produce the following minimum specifications: Yield point, 55,000 lbs. per sq. in.; tensile strength, 85,000 lbs. per sq. in.; elongation, 22 per cent; and reduction of area, 40 per cent. However, several of those combinations tend to produce machining difficulties under certain conditions.

When it is necessary for machine work to be done 3 in. or deeper

¹ Consultant on Steel Castings, Chicago.

² Wilkinsburg, Pa.

under the surface of castings of the chromium-manganese-molybdenum type, hard spots frequently are encountered. They are complex carbides of iron-chromium-manganese-molybdenum and probably are precipitated during the cooling of the casting in the mold. These spots occasionally attain the size of a 25-cent piece and are unmachinable with ordinary tools. The chromium-molybdenum and manganese-molybdenum occasionally show this condition, but to a lesser degree.

At the time this condition first was noticed, it was thought that improper operating conditions in the open hearth caused it. Considerable investigation lead us to believe that while inferior melting practice increased this trouble, good practice did not always eliminate it. Pouring temperatures next were considered. It was found that when castings weighing 1600 lbs. were poured in excess of 2825 degrees Fahr., hard spots frequently developed. Since it often was necessary to have 50 or more openings in a heat, it was impossible to pour all the castings at the proper temperature, and the chromium and manganese mixes were not used when castings were poured in which deep machine cuts were to be made. Nickel-molybdenum and vanadium-molybdenum mixes were free of the precipitated carbides. Prolonged soaking at temperatures in excess of 2000 degrees Fahr. dissipate them to some extent.

Have such hard spots been reported, and if so, to what are they attributed? The approximate analysis of the steels previously mentioned are shown in Table 1.

Table 1
Composition of Steels*

Type	Mn.	Cr.	Ni.	Mo.	V.
Cr-Mo70	.85	..	.35	..
Mn-Mo	1.3035	..
Cr-Mn-Mo	1.20	.75	..	.35	..
Ni-Mo70	..	1.35	.35	..
V-Mo7035	.12

* Carbon 0.30 per cent, silicon 0.30 per cent, sulphur and phosphorus 0.05 maximum.

H. W. GILLETT²: I have never encountered hard spots in molybdenum steels. Observations of myself and others indicate that such spots are not common in steels of that type.

I would like to have Mr. Critchett tell us a little more regarding nitrogen. Is its use advisable with the heat resistant alloys high in chromium and nickel? What are the limitations of its use?

J. H. CRITCHETT⁴: Nitrogen is used largely in castings where there is difficulty from large grain size. I don't think nickel-chromium steels are subject to that difficulty as much as the 25 per cent chromium steels. Therefore, the use has been largely in the plain chromium steel field. Nitrogen has some interesting effects on steel. It is far more effective than carbon in promoting a tendency toward austenitic structure. It gives a stiffening of the yield point without sacrifice of other good qualities, such as ductility and elongation and has a tendency to increase shock

² Battelle Memorial Institute, Columbus, O.

⁴ Union Carbide and Carbon Research Laboratories, New York.

resistance. The use of nitrogen does not seem to have any seriously deleterious effect. In 18-8 steel, an important effect is to improve machinability. Why, I do not know yet. A small addition of nitrogen to that steel will improve greatly the machinability.

There is a trick in the use of nitrogen. It has to be introduced into the bath at the proper time and the steel should not be overheated too severely. A little time should elapse between the addition of the nitrogen and casting. That enables the bath to reach equilibrium. A ferrochrome containing approximately 0.70 per cent nitrogen is the source of that element. It is used in the proportion that will give the desired analysis. Nitrogen in gates, risers and returned scrap is recoverable on remelting.

H. W. GILLETT: When you are using titanium and you also add nitrogen, do you form titanium nitride?

J. H. CRITCHETT: I have no information on that point. In general, titanium has been used in steels containing less than 25 per cent chromium. Most of our experience has been with the nitrogen addition only.

H. F. TIELKE*: What is the effect of carbon content on the higher chromium steel? Does the variation in chrome from 25 to as high as 31 per cent have any effect on heat and corrosion resistance?

J. H. CRITCHETT: Increasing the chromium above 25 per cent has a slightly beneficial effect both on the heat and corrosion resisting properties of plain chromium steels. A 30 per cent chromium steel has slightly better heat and corrosion resisting properties than the 25 per cent variety. On the other hand, the difference between the 35 and 30 per cent steels is not nearly as marked as the difference between those containing 25 and 20 per cent chromium. There seems to be a point in the range between 20 and 25 per cent chromium where the heat resisting properties, particularly, are much improved. The change occurs at approximately 22 per cent chromium. Then there is an almost straight line gradual improvement in properties from that point to 30 per cent chromium or more.

I do not think that there is as marked improvement in corrosion resistance as in heat resistance. That class of steel seems to be useful for resistance to corrosion mainly in the presence of mixed acids as in mine water and powder manufacture. For straight resistance to miscellaneous corroding media, I would choose another steel, some of the higher chromium or chromium-nickel steels. Apparently the properties at 25 per cent chromium are about as good as they are with 3 or 4 per cent more of that element.

The carbon effect does not seem to be as important in steels of that high chromium content as in steels containing less chromium. The control of carbon is vital in a 12 to 14 per cent chrome steel and also in the 18-8 variety. With 25 per cent chromium, the carbon exerts little effect on the resistance either to corrosion or oxidation. In steel castings, 0.40 or 0.50 per cent carbon seems to give good results. Similar carbon content in either of the other grades mentioned would be ruinous. Under the microscope, up to about 0.20 or 0.22 per cent carbon, nothing much is visible. The carbon is in solution and has no appreciable effect on the

*Crucible Steel Casting Co., Cleveland, O.

structure. Carbon is not as critical in the high chrome steel as in other corrosion resisting steels.

H. F. TIELKE: My experience has been that carbon is quite an important item. If the metal contains much over 0.50 per cent carbon, trouble may develop in certain applications. Castings working at 2100 degrees Fahr. have failed when the carbon content was between 0.65 and 0.70 per cent but gave good service with a carbon content of 0.40 to 0.50 per cent.

A. W. LORENZ⁶: Has Mr. Critchett any information on the addition of nitrogen to low alloy steels? There has been considerable mention recently in the technical press regarding the evil effects of nitrogen. For instance, in the development of the shielded-arc process of welding, one of the advantages claimed is that the shielded arc tends to eliminate nitrogen from the weld, making it much more ductile. We know that this is the case. A recent article in the technical press states that over 0.07 per cent nitrogen is extremely detrimental to steel. I think Dr. Westgren, in a letter to *Metal Progress*, mentioned one case where the accumulation of nitrogen in the scrap used in the basic electric process had increased the content in the bath finally to such a figure that the steel became brittle.

Have you done any work along those lines or has anyone else information regarding the accumulative effect of nitrogen? These data do not seem to conform with the information obtained on high chromium steels.

The written discussion inferred that steels with certain types of molybdenum additions were subject to hard spots. Our experience has been that when the machine shop complains of hard spots, the trouble is due to sand spots or dull tools. We have used forged nickel-chromium steels, with a high enough alloy content to be somewhat air-hardened, which passed through one machine and were machinable easily and then another machinist would complain that the steel contained hard spots.

On investigation, we found that it was a matter of dull tools with insufficient clearance. Dull tools worked on the material until a thin glaze of extremely hard steel was formed. It was possible to get under the hard spot with a chisel and chip out the glazed portion. I do not think hard spots in molybdenum steel often are due to any chemical constituent formed by the molybdenum but rather to some simple natural cause.

J. H. CRITCHETT: With respect to hard spots, another angle we have run into at times is that when the machinist complained of hard spots, the trouble was due to soft spots. Any difference in hardness between one grain and the next affects the action of the tool. In that case, the cutting edge may jump and glaze a short section. That section actually is softer than the surrounding metal. I agree that when hard spots are found, it is well to conduct a further examination to determine the exact cause of the trouble.

H. W. GILLETT: I think that Mr. Critchett has handled the question of hard spots very well. To have all the information, it may be well to

⁶ Bucyrus-Erie Co., South Milwaukee, Wis.

know whether molybdenum was added as molybdenum or as an oxide. My impression is that we should look first for extraneous causes before getting into complex chemistry.

C. E. SIMS⁷: This matter of hard spots is something pertinent to all alloy-steel castings. There are different causes for hard spots. I agree with Mr. Lorenz, and some of the other speakers, that many of the hard spots are due to sand inclusions, dull tools, or some mechanical cause. Nevertheless, there is a condition which produces hard spots in alloy steels of the type to which I think Mr. Frisbie refers. The explanation for this type of hard spot is simple enough. Theoretically the cure is as simple as the explanation, but practically the cure often is difficult. The micrographs of Fig. 3 in the paper by Mr. Strauss on vanadium steel can be used as illustrations in this discussion.

Although the common alloying elements, such as manganese, nickel, chromium, etc., are soluble largely in the solid state, during freezing there is considerable segregation. We are all more or less familiar with the freezing phenomena of an alloy such as steel. First, ribs of pure ferrite crystallize out, leaving the mother liquor richer in the impurities or alloying constituents until finally a eutectic is reached. The material that freezes later is continually richer in the alloying constituents. This gives a steel a mottled appearance, which is analogous to banding in rolled steel. This mottled appearance is shown in the micrographs of the normalized steels in Fig. 3 previously mentioned. Dark streaks represent areas that are richer in the particular alloying constituent present.

Under certain conditions, these richer areas reach a concentration of the alloying element that gives them air hardening properties. The relative concentration of the alloy will depend upon the general composition of the steel and the condition of freezing. In general, anything that furthers greater segregation, will tend to produce a greater differential between the richness of the alloying constituent in the various parts. In certain alloy steels, which in a normalized condition would have a general hardness of about 20 to 22 Rockwell C, certain small areas, easily distinguishable under the microscope, exist with a martensitic structure which would run as high as 45 to 50 Rockwell C. The steel in these areas is sufficiently high in alloys to be air hardening. The cooling rate has a great effect on these hard spots. Steel that shows pronounced hard spots in the normalized state will be free from them in the annealed state. If the cooling rate is accelerated somewhat above ordinary normalizing, they will be accentuated greatly. The cure is in the readjustment of the composition, a different proportioning of carbon and the alloying constituents.

Molybdenum might be a saving factor in this case. It might be possible to decrease the quantity of some other element and add a small amount of molybdenum. Reducing the concentration in this way would tend to eliminate the hard spots, while the addition of the molybdenum might give the same hardening characteristics as with the higher concentration of alloy element minus the molybdenum. Cases are known where

⁷ American Steel Foundries, East Chicago, Ind.

hard spots were eliminated by using two alloying constituents where only one had been used previously.

H. W. GILLETT: Your suggestion sounds as though you were talking about tiny spots. Mr. Frisbie stated he had hard spots as big as a quarter.

C. E. SIMS: The hard spots I referred to show up during machining as areas about 1/16 inch diameter.

JEROME STRAUSS⁸: I recall an article recently written by H. H. Morgan, Robert W. Hunt Co. on the occurrence of hard spots in manganese steel rails, in which he records segregations containing approximately 4 per cent manganese in a rail having an average content of this element of 1½ per cent.

Bearing on Mr. Sims' discussion and Dr. Gillett's question, within the past week we made two small melts in an Ajax furnace, under conditions insuring complete solution and thorough mixing, whose average composition was approximately 0.40 per cent carbon, 1.20 per cent manganese, and 0.70 per cent chromium, with other elements normal. The heat treatment involved a final tempering at 750 degrees Fahrenheit. Large areas of the test bars were almost unmachinable. While not making any systematic study of the segregation, we observed in the first heat a variation in manganese from 0.78 to 1.70 per cent and in the second from 0.99 to 2.81 per cent.

In respect to Mr. Lorenz's discussion on the influence of nitrogen, it is important that we regard its mode of combination and degree of dispersion. This applies particularly to the consideration of electric welds. That those made with bare electrodes can not but be lacking in toughness is evident from microscopic examination. At high magnification large brittle intercrystalline masses of nitride are evident which, when the nitrogen content is about one tenth of one per cent or more, approach a continuous network. Such material can obviously have little ductility. On the other hand, a low alloy steel containing nitrogen, distributed in the manner in which it occurs in the surface of nitrided steels, would be expected to show quite different properties.

CHAIRMAN R. A. BULL: Referring to the description of the particular test bars, was the segregation throughout the bar?

J. STRAUSS: The segregation appeared to be confined to the metal near the end and bottom surfaces of the castings and was greater at one end than at the other. It may be well to record here that the castings were poured in dry sand.

W. C. HAMILTON⁹: The subject of welding has not been brought up in this symposium, but it is important in connection with all castings. Considering two steels of the same approximate physical properties in a normalized condition, obtained from entirely different alloys, is it possible to weld one without any preheating or any heat treatment subsequently yet the other require strict precautions?

H. F. TIELKE: Some of the more complicated alloys such as tungsten, cobalt and chromium steels are extremely difficult to weld, even with preheating. The lower alloys present little difficulty in welding. They

⁸ Vanadium Corp. of America, Bridgeville, Pa.
⁹ American Steel Foundries, East Chicago, Ind.

should all be heat treated after welding regardless of composition.

P. E. MCKINNEY¹⁰: It would be unfortunate if we allowed this symposium to go into the records of the American Foundrymen's Association without having some contribution on some of the other influences of the various alloys mentioned. These include the effect on welding practice, fluidity, hot tears, the extent to which methods for equalizing the effect of nonuniform cross sections, such as by chilling, can be used, etc. The discussion would not be complete unless we had some idea of the relative casting properties, the extent to which various alloy compositions are susceptible and applicable in various cross-sections, considered from the standpoint of uniformity of design and other problems which every practical foundryman must solve when selecting an alloy to use for a complicated casting.

J. C. PENDLETON¹¹: I use alloy steel in all my casts. I do not use it for the strength but to eliminate hot tears. Certain alloys do that especially in extremely complicated casts. I use as high carbon as possible and nickel because I believe that it increases fluidity. I do not have any trouble getting the strength. My idea is simply to get the casting without a hot tear.

A. G. ZIMA¹²: In regard to the question of nickel and its influence on the fluidity of casting steel, I have found that quite a number of foundries have attributed their freedom from hot tear troubles and other foundry difficulties to increased fluidity imparted by small percentages of nickel. I have tried to secure specific data on this subject but have not been successful yet.

C. E. SIMS: Have either of the last two speakers any idea or explanation as to why increased fluidity helps to prevent hot tears?

P. E. MCKINNEY: If castings can be poured at a lower temperature, better results seem to be obtained. If most specifications would not restrict carbon content to 0.35 per cent, less hot tears would result.

J. STRAUSS: It seems to me that this question of hot tears is very closely related to a subject upon which we have very little information. Not that I am able to contribute any at this time, but perhaps a suggestion to well equipped universities seeking suitable problems for laboratory investigation might be effective. If we really knew something of the strength and ductility of these various cast steels at temperatures between the solidus and the lower transformation, we might be better fitted to rationally discuss this question.

F. A. MELMOTH¹³: I would like to know more about the methods of producing those steels. Were they made in the induction furnace, open hearth, etc.? I also would like to know a little about the type test bar from which the tests were cut and more details about the various heat treatments.

Our knowledge of alloy steel castings has not reached finality when we know how to produce certain tests from a prepared test bar made under highly controlled conditions. I agree with the suggestion that it

¹⁰ Bethlehem Steel Co., Bethlehem, Pa.

¹¹ Newport News Shipbuilding and Dry Dock Co., Newport News, Va.

¹² International Nickel Co., New York.

¹³ Detroit Steel Casting Co., Detroit.

might be studied further and as soon as possible arrangement should be made to get some information on the behavior of alloy steels in the foundry itself.

H. W. MAACK¹⁴: With reference to failure of steel castings by cracking, cracks in castings of an air hardening steel may be a quite different effect from that of hot tears in other steel compositions. We know from heat treating experience that if an alloy or high carbon steel is not drawn soon enough after it has been quenched, it may rupture from strains. I think we can expect the same of air hardening steel castings. They must be nursed along slowly through the cooling period to avoid cracking. It was brought out that nickel-vanadium steels are extremely shock resistant. Data are published to show that nickel-molybdenum steel also is highly shock resistant. I would like to ask Dr. Gillett if he has information on the possible beneficial effect of molybdenum in conferring impact resistance at subnormal temperatures.

H. W. GILLETT: I don't think you can credit molybdenum with any notably beneficial effect at low temperatures. Nickel is a specific for low temperature properties. In our own work, and that which was done at Wright Field, carbon steel seems to be the one that is susceptible peculiarly to low temperature. Most of the alloy steels that have been studied were far less susceptible. Of all the alloys, nickel apparently has the strongest effect in avoiding low temperature difficulties.

C. E. SIMS: You refer to a homely problem in the matter of spot welding. Any alloy steel that is spot welded, especially when arc welding is used, as in the correction of a small surface defect, will have a hard spot around this weld unless it is annealed subsequently. Often hard spots are encountered during machining, which are about the size of a quarter and in the shape of a ring. A weld will be found in the center of this ring.

GEORGE BATTY¹⁵: If the foundryman will apply some of these peculiar characteristics, he sometimes can make capital out of them. I have in mind a case where a bucket casting was not working well. The sockets of the teeth were worn badly. Knowing the peculiarity Mr. Sims has spoken of, we applied similar arc welding in spots to create these hard rings, and produced a local abrasion resistant material.

J. STRAUSS: I should like to refer briefly to the subject of mechanical properties at low temperatures. These comments are based on experiments with wrought steels at lower temperatures than have thus far been considered in this discussion; namely approximately 300 degrees Fahrenheit below zero. Carbon steels are extremely poor at such temperatures. The low alloy steels are only a little better. However, it was evident in tests made at the U. S. Naval Gun Factory, Washington, 7 or 8 years ago that the distribution of the micro-constituents might have some bearing as there were very appreciable differences in the degree of embrittlement of carbon steels depending upon the heat treatment employed. I think that some day we may have more complete information if time is afforded to give the problem ample attention.

¹⁴ Crane Co., Chicago.

¹⁵ Steel Castings Development Bureau, Philadelphia.

(Contributed as written discussion)—In Table 2 of the paper, "Chromium in Steel Castings," are shown the properties of a manganese-silicon-chromium-vanadium cast steel in several conditions of heat treatment including the normalized and tempered state. These properties do not compare favorably, particularly in respect to elongation and reduction of area, with those of the manganese-chromium-vanadium steels, Nos. 6 to 8 of Table 5, of my own paper presented at this session. Of these three steels, No. 8 compares closely in respect to composition, the major exception being the silicon content, which is 0.47 per cent as compared with Mr. Critchett's 0.80 per cent. While there are many variables that enter into the final properties of a test bar of cast steel in addition to composition, other experiences have led to the belief that high silicon content in low alloy steels, especially above 0.70 per cent, is not desirable and that other elements introduced to obtain equivalent hardness result in better values of ductility. From these tests, it is apparent that somewhat better results may be obtained by suitably balancing composition and heat treatment.

In Table 3, Mr. Critchett has given information on the properties of medium chromium content steels containing other elements. As a further contribution, the following four tests made in the laboratories of the Vanadium Corporation of America are quoted in Table 2.

Table 2

C.	Mn.	Si.	Cr.	W.	V.	Yield Pt.	Ult. Str.	% Elong. in 2 Inches.	% Red. in Area
						Lb./Sq. In.	Lb./Sq. In.		
.34	.55	.34	1.16	.92	.27	67300	101500	25.0	53.0
.32	.56	.33	1.28	2.25	.27	82800	115250	19.0	41.3
.26	.49	.17	1.23	4.50	.19	68150	100600	24.5	56.2
.19	.50	.24	5.17	37100	76950	34.0	78.4

All of these tests were taken from double normalized and tempered material. The properties of the three heat-resisting steels are excellent but it is of particular interest to note the high reduction of area of the low carbon plain chromium steel, which is unusual for a casting even of such low tensile strength. At the same time, the yield ratio is very low, probably, because the second normalizing temperature was too low for effective hardening. That could be improved without sacrifice of ductility by modification of composition.

J. H. CRITCHETT: In the early part of Mr. Strauss' written discussion, he brings up some difference between the properties of steel cited in my paper and an analogous steel based mainly on the silicon content and whether it is desirable to have over 0.70 per cent silicon. That depends a tremendous amount on the way the heat is made and the way the silicon is added. Some figures have been contributed that are even more favorable than those in my paper. In choosing the figures in my tables, I have not in any instance picked the highest that could be found. All the examples are taken from commercial production. Test bars and methods of testing are those which are standard in foundries.

Considerable work has been done lately in our own laboratories on

low temperature properties. In general, the best low temperature properties seem to be derived from a combination of alloying elements. When an austenite forming element is combined with a carbide forming element, better low temperature properties are obtained than when either is used alone. In general, alloy steels are much superior to plain carbon steels for low temperature properties.

P. E. MCKINNEY: Regarding the high silicon alloy mentioned, have you made any actual comparative studies on the amount of shrinkage as compared with manganese steel with about the same manganese and carbon content?

J. H. CRITCHETT: I have no definite data but from observations of castings made from the same pattern, there does not seem to be enough difference between the two alloys to create any new problem.

P. E. MCKINNEY: The reason I ask that is because we have observed that high silicon content produces much more internal shrinkage trouble.

J. STRAUSS: The work I had in mind showed a marked decrease in the ductility of a number of cast low-alloy steels when the silicon exceeded 0.70 per cent. Better ductility for given strength or yield point could be obtained by a combination of the carbon and alloying elements that did not involve or require silicon in excess of that amount.

A. G. ZIMA: One of the things I noticed in writing this paper on "Properties and Uses of Some Cast Nickel Alloy Steels" was the difficulty of obtaining consistent impact data. A foundryman wrote me recently that he had made a series of impact tests with a simple notch. The bars were heat treated identically but the notches were made in different manners. In one case, the notch was drilled, in another case drilled and subsequently machined, but to identical dimensions. He claimed that he got greatly varying results by these methods of making the impact specimen. Not only that, but we find impact specimen of various dimensions and it is difficult to obtain any consistent or relative figures. It might be a good move for the foundrymen to get together on one or possibly two standard impact tests and develop a procedure that would be comprehensible throughout the industry.

In regard to hard spots, about four or five years ago, I visited a foundry annealing about 5,000 small pinions made of nickel-chromium-molybdenum, air hardening steel. They were in a hurry and dumped the gears upon the furnace car in a large heap, pushed the car into the furnace, heated it the prescribed length of time at the proper temperature, removed it, and sent the castings to the machine shop. Brinell hardness ran from 200 to 400. Hardly any two gears were alike. The reason was that castings at the top of the pile obtained thorough annealing while some in the center never reached the critical point. Various parts of the charge were receiving different treatments, although the pyrometer indicated that the charge was annealed properly. Another reason may be welding or spotting of pin holes and small imperfections in the casting,

or removal of the risers with the acetylene torch without subsequent annealing. All of those contribute to hard spots.

I have an idea regarding the large crystals which Mr. Critchett mentioned on high chromium steels. There is an excellent contribution on the control of grain size by the control of pouring temperature. This was written by Rys in *Kruppsche Monatschafte*, April, 1930, page 50.

Silal and Nicrosilal Heat-Resisting Cast Irons

By A. L. NORBURY,* D.Sc., AND E. MORGAN,* M.Sc.,
BIRMINGHAM, ENGLAND

Abstract

The present paper contains a general account of the properties of two recently developed heat-resisting gray cast irons. Detailed accounts of their mechanical and physical properties, microstructures, etc., have been previously published (see footnotes 1, 2). These irons, under the names of "Silal" and "Nicrosilal," are high-silicon irons with nickel and chromium in the Nicrosilal. Effects of the various elements are discussed. Very definite advantages in regard to scale and growth resistance are cited.

1. *Silal*¹ is a gray cast iron containing for most purposes about 5 per cent silicon and 2.5 per cent total carbon, and having a fine graphite structure. The increased silicon content greatly increases the resistance to oxidation, and the iron gives much better service than the best low-silicon irons for purposes such as fire-bars, furnace parts, etc., at temperatures up to about 850 degs. Cent. (1562 degs. Fahr.), as illustrated in Figs. 1 and 2.

2. The iron, however, is not suitable for castings which are

* British Cast Iron Research Association.

¹ Norbury, A. L., and Morgan, E. "The Effect of Carbon and Silicon on Growth and Scaling of Gray Cast Iron." *Iron and Steel Inst. Journal*, 1931, n. 1, p. 413.

British Patent 323,076 of 1929.

Norbury, A. L., and Morgan, E. "The Effect of Melting Conditions on Microstructure and Mechanical Properties of Gray Cast Irons Containing Various Amounts of Carbon and Silicon." *Iron and Steel Inst. Journal*, 1930, n. 1, p. 367.

² Norbury, A. L., and Morgan, E., "Nickel-Chromium-Silicon Cast Irons." *Iron and Steel Inst. Journal*, 1932, n. 2.

Norbury, A. L., and Morgan, E., "Manganese-Silicon and Nickel-Silicon Cast Irons." *Iron and Steel Inst. Journal*, 1932, n. 1, p. 393.

British Patent 378,508 of 1932.

NOTE: This paper was presented before one of the sessions on Cast Iron at the 1933 Convention of American Foundrymen's Association.

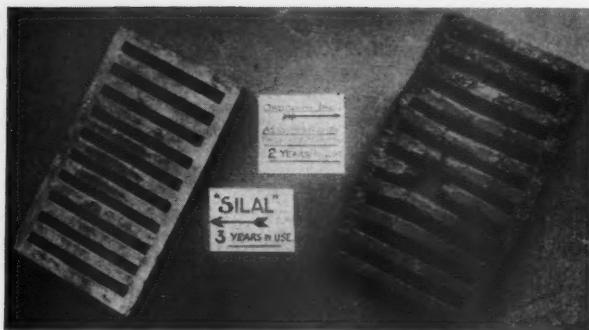


FIG. 1—Left: SILAL FIRE GRATE AFTER THREE YEARS IN SERVICE. Right: FIRE GRATE OF ORDINARY IRON AS SUPPLIED BY FIREPLACE MANUFACTURERS, AFTER TWO YEARS IN SERVICE. (PHOTOGRAPHS ONE-THIRTEENTH ACTUAL SIZE, SUPPLIED BY SMITH, PATTERSON & CO. LTD., BLAYDON-ON-TYNE.)

subjected to rapid local heating or cooling, since under such conditions it is liable to crack owing to its brittle character. It is easily machinable and can be produced in the cupola.

3. *Nicrosilal* is an austenitic gray cast iron, similar in certain respects to Niresist,² containing for most purposes about 2 per cent total carbon, 5 per cent silicon, 1 per cent manganese, 18 per cent nickel and 2 to 5 per cent chromium. The iron is extremely tough and ductile, in spite of its high silicon content, which increases the resistance to oxidation as in the case of Silal. The oxidation resistance is further increased by the chromium content, so that even at 950 degs. Cent. (1742 degs. Fahr.), sealing only proceeds very slowly (see Fig. 3), and growth is practically nil.

4. The iron is easily machinable and takes a very fine finish. It is non-magnetic and, like Niresist, has good corrosion-resisting properties.

5. Certain compositions may be cast white and malleableized by heating for half an hour at 950 degs. Cent. (1742 degs. Fahr.), as shown in Fig. 4; others may be hardened by tempering at about 600 degs. Cent. (1112 degs. Fahr.). The iron can be produced in the cupola and has good fluidity (see Fig. 4) and "castability."

PROPERTIES AND COMPOSITION OF SILAL

Silicon Content.

6. The high silicon content of Silal has several advantageous effects on the heat-resisting properties. It increases the resistance

² Vanick, J. S., and Merica, P. D., "Corrosion and Heat-Resistant Nickel-Copper-Chromium Cast Iron." *Transactions A.S.S.T.*, 1930, v. 18, p. 923.

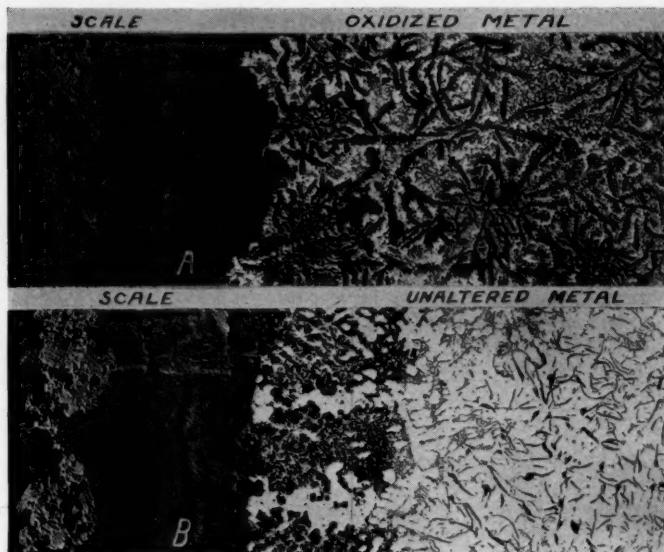


FIG. 2—*A*: EDGE OF FURNACE-PLATE CASTING, APPROXIMATELY 3 FT. x 2 FT. x $\frac{1}{2}$ IN.), AFTER THREE FIRINGS AT 850 DEGS. CENT. (1562 DEGS. FAHR.) CASTING WAS OXIDIZED THROUGHOUT, DISTORTED AND CRACKED, LENGTH INCREASING APPROXIMATELY $1\frac{1}{4}$ INCHES. X75. *B*: EDGE OF SIMILAR CASTING IN SILAL AFTER 60 TO 70 FIRINGS AT 850 DEGS. CENT. (1562 DEGS. FAHR.) CASTING STILL IN GOOD CONDITION, COATED WITH THIN ADHERENT SCALE, LENGTH INCREASING APPROXIMATELY $\frac{1}{2}$ INCH. X75; REDUCED ONE-THIRD IN PRINTING.

to oxidation very considerably (see Figs. 1 and 2); it raises the temperature of the discontinuous volume change which accelerates growth to above 900 degs. Cent. (1652 degs. Fahr.), and consequently to above the temperature reached in service; it lowers the total carbon content; it increases the strength at the working temperature, and it eliminates combined carbon which may graphitize and cause a volume increase.

7. On the other hand, the high silicon content has the detrimental effect of making the iron brittle and low in impact strength and ductility at temperatures below a black heat. At higher temperatures it is quite ductile and does not crack, and is considerably stronger than a low silicon iron.

8. For most purposes a silicon content of about 5 per cent has proved best, since such an iron is easily machinable, not too brittle, and has a tensile strength of up to 44,800 lbs. per sq. in., with a total carbon content of 2.5 per cent. Such a composition

can easily be made in a cupola from ferrosilicon and steel scrap, since the solubility at a given temperature of carbon in molten cast iron is lowered about 0.25 per cent total carbon by each 1 per cent silicon present.

9. A total carbon content of about 2.5 per cent gives the best compromise of properties for light castings; although a lower total carbon content gives greater heat-resistance and strength, it increases the liquid shrinkage, lowers the fluidity, and increases the tendency to mottle on thin sections. For heavier castings the total carbon content should be lowered in order to obtain as fine a graphite structure and as high a strength as possible.

Phosphorus, Manganese and Sulphur Contents.

10. The phosphorus content has the same effect that it has on low-silicon irons, and is best at not more than about 0.5 per cent. The manganese should not be greatly in excess of that required to neutralize whatever sulphur is present, which requires about $(0.3 + 1.7 \times \%S)$ per cent manganese.⁴ If the manganese is much higher than this it tends to make the iron pearlitic, especially if it is cooled rapidly, which reduces the machinability. Moreover, manganese decreases the resistance to scaling.

Oxidation Resistance of 5% and 10% Silicon Silal Irons.

11. The 5 per cent silicon iron gives good resistance to growth and scaling at temperatures up to about 850 degs. Cent. (1562 degs. Fahr.), but at higher temperatures it scales more rapidly and, unless the total carbon content is very low and the graphite flakes very thin, may give worse service than a low-silicon iron. Being more brittle, the growth more readily produces cracks down which the oxidation advances toward the interior.

12. This is in agreement with Carpenter and Rigan's finding⁵ that 4 to 6 per cent silicon irons, which had very coarse graphite structures, grew more rapidly and to a greater extent than low-silicon irons when repeatedly heated at 950 degs. Cent. (1742 degs. Fahr.) in a muffle. If, however, the silicon content is increased, the resistance to scaling is also increased and with, for example, about 10 per cent silicon present, scaling is very slow at 950 degs. Cent. (1742 degs. Fahr.)

13. The practical application of a 10 per cent silicon iron,

⁴ Norbury, A. L., "Manganese in Cast Iron." *Inst. of British Foundrymen Proceedings*, 1928-29, v. 22, p. 151.

⁵ Carpenter, H. C. H., and Rigan, "The Growth of Cast Irons After Repeated Heating." *Iron and Steel Inst. Journal*, 1909, n. 2, p. 29; and 1911, n. 1, p. 196.

however, is more limited on account of its increased brittleness. In such an iron the total carbon content must be further reduced, to about 1.6 per cent for light castings, in order to obtain a fine graphite structure.

PROPERTIES AND COMPOSITION OF NICROSILAL

General Properties of Nicrosilal.

14. Nicrosilal,² which contains for most purposes about 2 per cent total carbon, 5 per cent silicon, 1 per cent manganese, 2 to 5 per cent chromium and 18 per cent nickel, has the good heat-resisting properties of Silal, on account of its high silicon content, together with further heat-resistance conferred by its chromium content (see Fig. 3) and toughness and ductility (see Fig. 4) conferred by its nickel content. It has not the undesirable brittle quality of Silal and is more resistant to oxidation and has greater strength at high temperatures.

15. It is remarkable that although the presence of 3 or 4 per cent silicon in pearlitic or ferritic irons reduces the ductility and elongation in tensile almost to zero, yet as much as 7 per cent silicon may be present in austenitic irons and excellent ductility still obtained, as indicated by a 3 per cent elongation in tensile. Tensile strengths of over 44,800 lbs. per sq. in. may be obtained and Brinell numbers as low as 110.

16. The iron is non-magnetic and has an electrical resistivity in the neighborhood of 150 micromhos per cu. cm., depending on the analysis. The average coefficient of expansion up to 1000 degs. Cent. (1832 degs. Fahr.) is about 1.8×10^{-5} per degree Centigrade, as compared with the 1.4 value of pearlitic and ferritic irons. It has good corrosion-resisting properties, and certain compositions may be malleableized by casting white and annealing for half an hour at 950 degs. Cent. (1742 degs. Fahr.)—see Fig. 4—while others may be hardened by tempering.

17. The iron can be produced in the cupola and has good fluidity and "castability" (see Fig. 4). Its freezing point is about the same as that of low phosphorus gray iron, but it requires more feeding on account of its lower graphite content plus higher casting temperature.

Total Carbon Content.

18. A total carbon content of about 2.0 per cent gives the best compromise of properties for light castings. There is no difficulty in obtaining such a total carbon from a cupola melt,

since the solubility of carbon in molten cast iron at a given temperature is reduced about 1.25 per cent by 5 per cent silicon, and about 0.7 per cent by 18 per cent nickel. The lower the carbon content, the greater the strength and heat-resistance, but the greater the liquid shrinkage and tendency to mottle on thin sections.

19. Such sections may be cast white by using a sufficiently low total carbon content (which is about 1.8 per cent in the 1/10 inch thick castings shown in Fig. 4) and then malleableizing by heating for about half an hour at 950 degs. Cent. (1742 degs. Fahr.), whereby a ductility is obtained comparable with that of malleable iron. White iron produced by a raised chromium content does not graphitize in this way.

Silicon Content.

20. For most purposes a silicon content of about 5 per cent is best, since compositions containing more than about 5 per cent silicon and less than about 17 per cent nickel have the property of hardening on tempering at about 600 degs. Cent. (1112 degs. Fahr.) This hardening is due to the conversion of austenite into

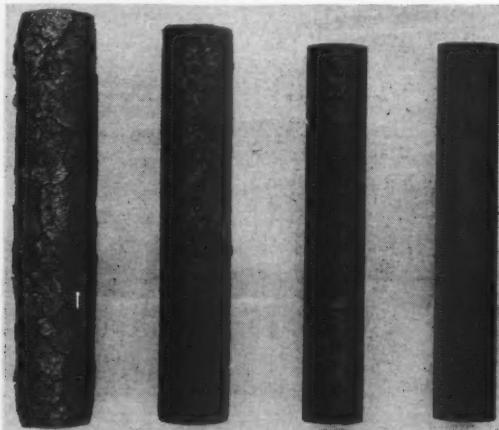


FIG. 3—EFFECT OF TEN HEATS, EACH OF THREE HOURS DURATION, ON 3 X 1/2 INCH BARS AT 950 DEGS. CENT. (1742 DEGS. FAHR.) A: HIGH-QUALITY ENGINEERING IRON, INCREASED 70% IN VOLUME, 21% IN WEIGHT. B: HIGH-QUALITY ENGINEERING IRON, INCREASED 47% IN VOLUME, 11% IN WEIGHT. C: LOW-CHROMIUM NICKOSILAL, DECREASED 2.5% IN VOLUME, 5% IN WEIGHT. D: HIGH-CHROMIUM NICKOSILAL, INCREASED 0.4% IN VOLUME, DECREASED 1.3% IN WEIGHT. (PHOTOGRAPHS AND TESTS SUPPLIED BY VICKERS-ARMSTRONG LTD., BARROW-IN-FURNESS.)

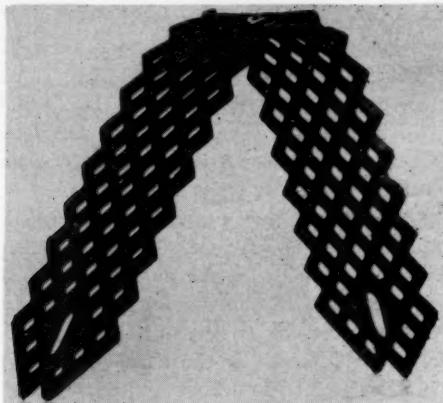


FIG. 4.—NICROSIL CASTING HAVING SECTION OF APPROXIMATELY 1/10 INCH, RUN WITHOUT DIFFICULTY IN GREEN SAND. CASTING WAS MOTTLED AS CAST, BUT WAS MALLEABLEIZED BY HEATING TO 950 DEGS. CENT. (1742 DEGS. FAHR.) ONE-HALF HOUR, AND WHEN COLD WAS BENT ALMOST DOUBLE.

martensite, and when it occurs the iron increases in volume and becomes magnetic.

21. The amount of austenite in a given iron to change in this way increases as the silicon content increases and as the nickel content decreases. If the composition and conditions of tempering are such that a considerable amount of austenite decomposes and the volume change is of considerable amount, it is liable to cause distortion, cracking and growth, especially if repeatedly undergone.

22. Consequently, for service conditions which involve repeated heating at about 500 to 700 degs. Cent. (932 to 1292 degs. Fahr.) for any length of time, compositions should be used which do not decompose to martensite to any great extent. On the other hand, compositions which can be hardened by tempering have interesting possibilities for applications involving wear resistance.

23. Another limit to the silicon content is that it should not exceed 6 per cent with 20 per cent nickel, or 7 per cent with 16 per cent nickel, since silicon in excess of the above amounts is not retained in solid solution but is present in the microstructure as a hard, brittle silicide which reduces the ductility and machinability. Apart from the above, the silicon should be as high as possible, since the higher it is, the greater the resistance to oxidation and the greater the strength at high temperatures.

Nickel Content.

24. The nickel content should be not less than 17 per cent if the above hardening on tempering is undesirable. If, however, the service conditions are such that the casting is not subjected to tempering at 500 to 700 degs. Cent. (932 to 1292 degs Fahr.) for any appreciable time, the nickel content may be as low as 13 per cent and machinable austenitic irons still obtained.

25. The above nickel percentages all refer to irons having chromium contents in excess of about 1.8 per cent. Although 18 per cent nickel is required to make an iron austenitic when no chromium is present, this 18 per cent is progressively lowered to 13 per cent nickel as the chromium content is increased to about 1.8 per cent; further increase in the chromium content does not further reduce the amount of nickel required.

26. If the nickel content is lowered below 13 per cent, hard martensitic irons are obtained which become progressively less ductile and give increasingly large volume changes on heating and cooling as the nickel is decreased. Finally, the irons become ferritic and brittle when the nickel is below about 7 per cent with 5 per cent silicon, 8 per cent with 6 per cent silicon, and 10 per cent with 7 per cent silicon. The lower limit of nickel for nearly all heat-resisting applications is, consequently, about 13 per cent. The higher limit may be as much as 20 per cent in cases where, for example, the surface of the casting is peened by cold chiseling which causes work hardening.

Chromium Content.

27. The chromium content should not be less than about 1.8 per cent, as indicated above, and should be as high as possible in order to obtain the greatest heat resistance, since chromium increases the resistance to oxidation and refines the graphite structure. Chromium may be increased for this purpose to a greater amount in the case of thick-section castings.

28. On the other hand, its carbide-forming action increases the tendency to mottle, increases liquid shrinkage, hardens the iron and reduces machinability. With about 1.8 per cent chromium the irons are easy to machine; with 5 per cent chromium the irons are becoming definitely mottled but are still machinable with high-speed tool steel.

Manganese, Phosphorus and Sulphur Contents.

29. The manganese content should be sufficient to neutralize

whatever sulphur is present, but it should not be much above about 1 per cent. Although it acts in the same manner as nickel in making cast iron austenitic (being about twice as powerful weight for weight), it has the undesirable action of reducing the resistance to scaling and of increasing the tendency to mottle.

30. The phosphorus content should be below 0.1 per cent if the iron is to be used at temperatures near 960 degs. Cent. (1760 degs. Fahr.), at which temperature any phosphide present in the iron melts and reduces the strength to a very low figure. Sulphur is undesirable, especially if it segregates, and it should be as low as possible.

DISCUSSION

NOTE: As the authors of the paper were unable to be present at the meeting, their paper was read by J. S. Vanick, International Nickel Co.

W. W. KERLIN:¹ The suggestion that this iron must not be heated too rapidly is a little bit hazy in my mind. My idea of a rapidly heated iron is an ingot mold for which Silal would obviously be unsuited. Would it be suitable for such a casting as an aluminum melting pot, which sometimes suffers from abuse in severe local heating?

J. S. VANICK: Of course, I plan to refer these questions to the authors ultimately, and will merely reply to any discussion in the light

¹ Gray Iron Institute, Cleveland, Ohio.

of any experience we may have had with similar compositions. My thought in the case of the aluminum melting pot would be, first; that perhaps the solubility of the silicon in the aluminum is so great that you would not have every good foundation material to work with; second, a possible high expansivity and low heat conductivity. You have already presumed that the Silal composition is not suitable, and I am inclined to agree, because aluminum pots, like many other castings, are heated in jerky cycles. I would not expect the Nicrosilal to be better because of the high expansivity and the risk of fire-cracking. High expansivity is usually associated with a somewhat lesser heat conductivity which accentuates the surface cracking effect.

MEMBER: I have had very little experience with Nicrosilal but we have done quite a lot with Silal. It is not quite as simple to handle as it first appears. The mixture can be melted easily and readily—there is no doubt about that—and you can get the very low carbons without any trouble, but the results depend entirely on the structure of the metal in the castings.

I have here some figures that were taken by an independent firm, which had an idea of using this metal for low temperature carbonization. There were four lots in the experiments. For one lot the total carbon was 2.59 per cent, silicon 5.85, manganese 0.91, sulphur 0.1, phosphorus .05, tensile strength 20,000 lbs. per sq. in. After heating to 800 or 900 degrees Cent. for eight-hour periods, the material being packed in fine coal, it was found after twenty-one heats that the linear expansion was 6.11 per cent over the original. The second lot had a total carbon content of 2.41 per cent, silicon 5.5, manganese 1.3, sulphur 0.03, and phosphorus 0.37. After twenty-eight heats this material had increased 2.75 per cent under the same conditions as the first lot.

What these people actually did was to take the linear expansion of heating to 800 to 900 degrees Cent. for eight-hour periods with fine coal, so as to get the SO_2 effect. The growth on the first lot of material, which was cupola melted, after twenty-one heatings of this eight-hour period, was a linear expansion of 6.11 per cent. The crucible melted material of the second lot, after twenty-eight heats only expanded 2.75 per cent, which is exceptionally close. I want to bring out particularly that while there is no doubt that this material as to its composition may be quite right the structure must be such that it produces the desired results. I have advanced this thought because some of you who may try this material may have it fail because of lack of the proper structure. The material, however, can be made to give good service.

MEMBER: This chromium ferrite, commonly called constituent B, is very brittle. Could not that be toughened up by heat treating? Have you had any experience along this line?

J. S. VANICK: I rather object to the designation chromium ferrite to begin with. I don't know how you are going to get those two elements together with a lot of carbon available, unless the silicon present is there in sufficient quantity to keep them from combining. If you heat treat or anneal any high chromium iron composition with carbon available you

may actually form and precipitate chromium carbide. There would be a tendency for chromium to leave the body or matrix and form a carbide very similar to what it does in stainless steels. I hardly expect a chromium ferrite to occur with so much carbon there, unless you consider the mass of silicon that is in with it.

Results of Comparative Tests of Titanium Treated and Other Alloy Cast Irons

BY G. F. COMSTOCK,* NIAGARA FALLS, N. Y.

Abstract

This paper presents test results on the effect of titanium in cast iron. Four series of tests were conducted, one of cupola-melted iron and three of electric-melted iron. Varying percentages of titanium were added to each series, as well as chromium, nickel and molybdenum. Check tests were run on irons without alloying additions. Physical properties of the resulting irons are reported, together with machining test results. The results lead to the conclusion that a high-strength cast iron may be made with better machinability and at less cost with titanium and chromium than with other alloys.

1. A new titanium alloy has recently been developed that is readily soluble in cast iron, and it is the purpose of this paper to show what results have been obtained from its use. The alloy has low contents of both carbon and aluminum and contains about 15 to 20 per cent each of titanium and silicon, the balance being iron. It will dissolve in small hand-ladles of cast iron at ordinary foundry temperatures, but the recommended method of use for the best efficiency is the same as for other alloys, or by dropping the alloy gradually in the stream of iron just as it leaves the furnace.

PRELIMINARY TESTS

2. Preliminary tests at several foundries with alloy additions of from 0.5 to 2 per cent showed that the chief effect of

* Metallurgist, Titanium Alloy Mfg. Co.

NOTE: This paper was presented before one of the sessions on Cast Iron at the 1933 Convention of the American Foundrymen's Association.

titanium was to decrease the size of the graphite flakes, and that there was also a tendency to decrease the content of combined carbon and increase the formation of graphite and ferrite. The finer graphite in the titanium-treated iron generally caused an increase in strength as compared with the same iron untreated, the improvement in strength being as much as 25 per cent in some instances.

3. That this effect was due to the titanium content of the alloy, and not the silicon content, was shown by testing iron treated with corresponding amounts of ferrosilicon, when it was found that silicon alone produced coarser graphite and decreased strength. When too much of the titanium alloy was added, however, so that the graphitizing effect was strong, there was too much ferrite in the titanium-treated iron, and the increase in strength expected from the refinement of the graphite flakes was lost.

Use of Hardener

4. This led to the idea of using a hardener such as chromium with the titanium alloy, to prevent ferrite formation, and allow the maximum improvement to be obtained from the finer graphite. Results in this direction were encouraging and several series of tests were made, in each of which the intention was to secure results from an iron of a definite constant composition untreated, treated with titanium, with chromium, with chromium and titanium, with chromium and nickel, and with molybdenum, respectively. This was best accomplished with electric-melted iron; in cupola practice it seemed to be difficult to get a series of irons with sufficiently constant silicon and carbon contents to permit accurate comparisons of the effects of the various alloys.

Tests on Cupola and Electric-Furnace Melted Metal

5. Results from four series of tests are presented in Tables 1 and 2. These include all the trials in electric-melted iron that have been made, and the one trial in cupola iron that showed fair constancy of carbon and silicon contents throughout the series of tests.

Methods of Testing.

6. The transverse test results are averages of from six to eight test bars, with not over two rejections on account of surface flaws. These values were corrected for size of bar to a standard diameter of 1.25 in. The bars were broken at slow speed, on sup-

Table 1
CHEMICAL, MECHANICAL AND CUTTING TESTS OF ALLOY CAST IRON

	Series A, Cupola Melted					Series B, Electric Melted				
	None	1	0.5	None	None	1	0.5	None	0.5	None
Ferrotitanium added, %	None	None	None	None	None	None	None	None	None	None
Titanium content, %	0.042	0.102	0.042	0.042	0.042	0.051	0.051	0.051	0.051	0.051
Chromium content, %	0.32	0.30	0.41	0.39	0.39	0.39	0.39	0.39
Other alloy content, %	0.94	0.94	0.94	0.94	0.94	0.94
Silicon content, %	1.78	1.78	1.68	1.88	1.91	1.94	1.96	2.00	2.23	2.21
Graphite content, %	2.99	3.03	2.85	2.83	2.92	2.98	2.99	2.91	2.96	2.74
Combined carbon, %	0.50	0.49	0.44	0.67	0.60	0.56	0.49	0.56	0.51	0.73
Total carbon, %	3.40	3.52	3.49	3.50	3.52	3.54	3.48	3.46	3.47	3.43
Transverse strength, lbs.	4,115	4,500	4,340	4,628	4,560	4,448	4,428	4,450	4,673	4,823
Increase over untreated iron, %	9.4	5.5	12.5	10.8	8.1	7.6	7.6	7.6	5.0	8.4
Deflection, inches	0.132	0.143	0.138	0.133	0.132	0.137	0.144	0.153	0.149	0.146
Tensile strength, lbs. per sq. in.	30,860	32,460	33,790	35,240	35,280	32,400	31,000	33,820	35,660	36,320
Increase over untreated iron, %	5.2	9.5	14.2	14.3	14.3	5.0	0.5	5.4	8.0	14.4
Brinell hardness at center	183	183	183	192	196	187	179	187	186	202
Brinell midway between center and edge	187	187	192	196	207	179	183	187	196	207
Comparative resistance to cutting:										
Shaper test, 0.01 in. depth	53	55	53	55	55	55	55	55	55	55
Shaper test, 0.02 in. depth	85	86	84	87	86	86	86	86	86	86
Lathe test, 0.0013 in. feed	27.5	28.5	25	25.5	25	26.5	26.5	26.5	22.5	22.5
Lathe test, 0.00694 in. feed	39.5	43.0	38	39.5	38	39.5	39.5	39.5	37.5	37.5
Check Lathe test, 0.00313 in. feed	23	26.5	22.5	24	22.5	24	24	24	24	24
Check Lathe test, 0.00694 in. feed	36	40	30	39	37	39	37	39	37	37
Drill test, torque	18	17.5	18	18.5	17.5	18.5	17.5	17.5	19.0	19.0
Drill test, thrust	850	840	870	820	800	800	800	800	880	880

Table 2
CHEMICAL AND MECHANICAL TESTS OF ALLOY CAST IRON

	Series C, Electric Melted					Series D, Electric Melted				
	None	1	1.5	None	None	None	1	1.5	None	None
Ferrotitanium added, %	0.042	0.114	0.126	0.138	0.138	0.020	0.054	0.098	0.143	0.051
Titanium content, %	0.72	0.74	0.73	0.75	0.75	0.42	0.34	0.38	0.43	0.36
Chromium content, %	1.27	Ni 0.42 Mo
Other alloy content, %	2.24	2.24	1.61	1.64	1.65	1.2 Ni
Silicon content, %	2.24	2.26	2.23	2.25	2.24	2.24	2.56	2.43	2.42	1.64
Graphite content, %	2.49	2.54	2.36	2.34	2.34	2.36	2.47	2.59	2.43	1.62
Combined carbon, %	0.68	0.64	0.83	0.82	0.84	0.81	0.69	0.61	0.60	2.46
Total carbon, %	3.17	3.18	3.18	3.16	3.18	3.17	3.16	3.20	3.16	2.48
Transverse strength, lbs.	4,400	4,512	4,380	4,980	4,625	4,672	4,382	4,092	4,350	4,920
Increase over untreated iron, %	2.6	12.0	5.1	6.2	6.3	3.7	4,956
Deflection, inches	0.125	0.134	0.107	0.112	0.110	0.100	0.115	0.108	0.110	15.0
Tensile strength, lbs. per sq. in.	38,840	39,900	43,020	46,680	45,980	42,940	44,300	37,600	39,600	40,440
Increase over untreated iron, %	2.7	10.8	20.2	18.4	10.5	14.0	11.0	11.0	12.3
Brinell hardness at center	192	196	217	223	217	217	207	212	217	228
Brinell midway between center and edge	196	207	223	235	223	228	212	207	217	223

ports 12 in. apart, with the beam automatically balanced; and the deflections were read from autographic stress-strain diagrams.

7. The tensile test results are averages from five tests made on the lower (as cast) halves of the broken transverse test specimens, with no rejections for flaws. These test-bars were machined to 0.8 in. diameter, with threaded ends, and pulled slowly with spherical-seated grips on the same machine with automatically-balanced beam.

8. The Brinell hardness tests were made on cross-sections filed and smoothed with coarse emery, using a 10 mm. ball applied with 3000 kg.m. for 30 seconds.

9. The tests for resistance to cutting were made by O. W. Boston of the University of Michigan, using the tool dynamometer described recently in *Metal Progress*. The values given are merely readings of his instrument and show the comparative resistance to cutting of the various samples under identical conditions. Samples for which no values are given were not tested.

Discussion of Test Results

10. On account of the restricted space allotted to this paper, no discussion of the tabulated results is possible. The microstructures and fractures of all the samples were studied, but only a very few typical illustrations can be given here in Figs. 1 and 2.

11. The structures showed very consistently finer graphite in the titanium-treated samples than in any of the others. In the iron treated with chromium alone, the graphite was a little coarser than in the titanium-treated iron. In the samples containing nickel or molybdenum, the graphite usually was much coarser or about the same as in the untreated iron.

12. The increased strength in the chromium-bearing irons seemed to be due to the absence of any free ferrite and to a finer lamellar structure of the pearlite. Chromium also was found to be evident in the microstructure because of the presence of a hard, white constituent, probably chromium-carbide, associated with the spotted steadite (phosphide eutectic) found in all the samples.

13. The molybdenum-treated iron did not show fine graphite, or finely lamellar pearlite, or absence of free ferrite even in Series B where its strength was high. The structural reason for the strength of this molybdenum-treated iron is, therefore, a mystery.

14. The fractures of these irons, both in the test-bar size and in $1\frac{1}{4}$ -in sections, also were interesting. The titanium-treated irons

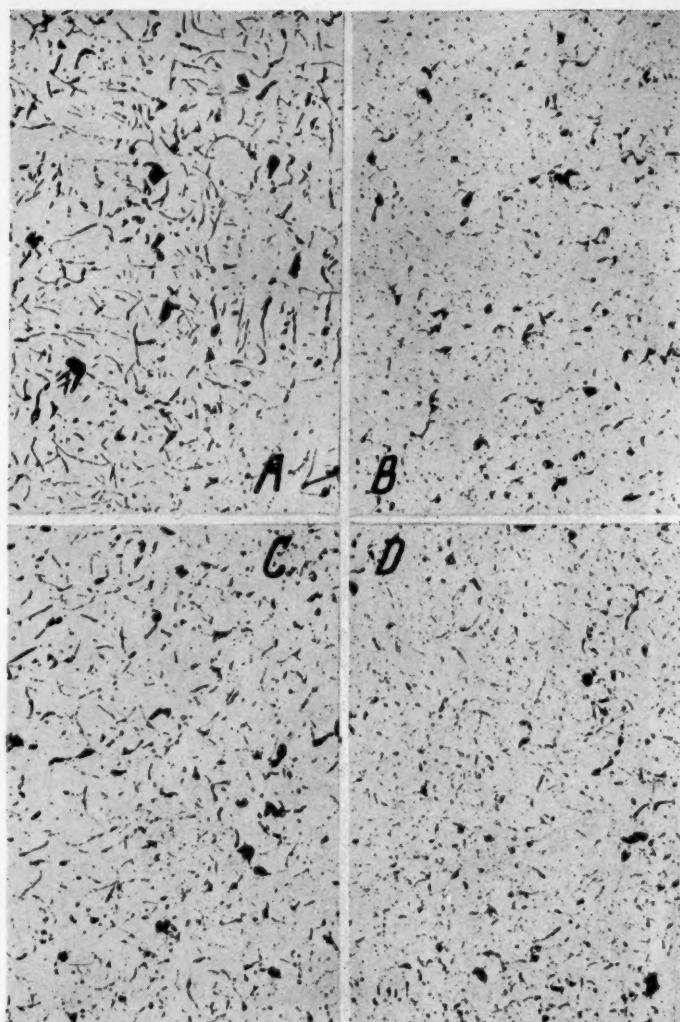


FIG. 1—EXAMPLES OF TYPICAL GRAPHITE PARTICLES ABOUT MIDWAY BETWEEN CENTER AND EDGE OF A CROSS-SECTION OF 1.25-INCH DIAMETER BAR FROM SERIES C (TABLE 2). X50, UNETCHED. A: PLAIN IRON, UNTREATED. B: IRON TREATED WITH 1 PER CENT FERROTITANIUM ONLY. C: IRON CONTAINING 0.72 PER CENT CHROMIUM, NOT TREATED WITH TITANIUM. D: IRON CONTAINING 0.74 PER CENT CHROMIUM, AND TREATED WITH 1 PER CENT FERROTITANIUM.

could be recognized in every series by their finer grain and darker fracture. The samples treated with chromium alone showed the most chill in the $\frac{1}{4}$ -in sections, and titanium was found to have the same effect as nickel in counteracting the extra chill due to

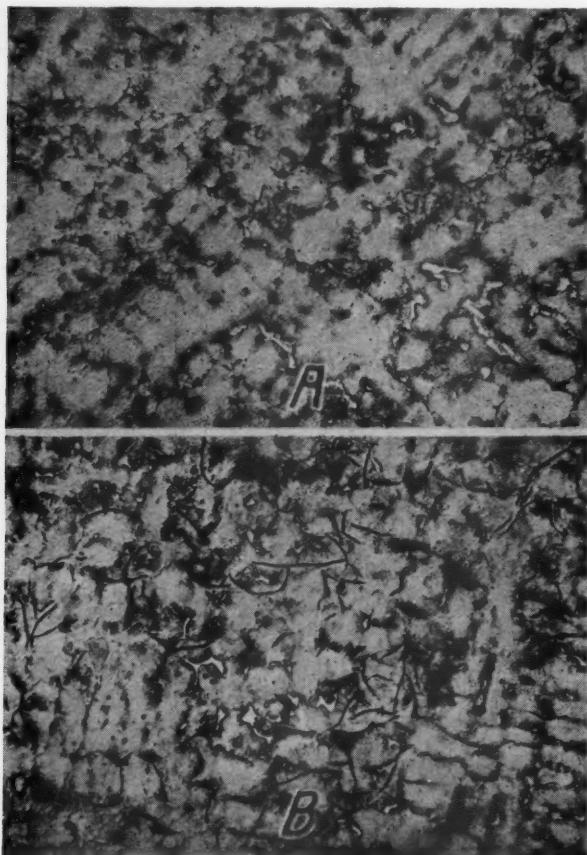


FIG. 2—EXAMPLES OF TYPICAL MICROSTRUCTURE ABOUT MIDWAY BETWEEN CENTER AND EDGE OF A CROSS-SECTION OF 1.25-INCH DIAMETER BAR FROM SERIES D (TABLE 2), X100, ETCHED WITH PICRIC ACID. BLACK SPOTS ARE GRAPHITE; GRAY IS PEARLITE OR SORBITE; BRIGHT AREAS ARE STEADITE, PROBABLY CONTAINING CHROMIUM CARBIDE. A: IRON CONTAINING 0.43 PER CENT CHROMIUM AND NO NICKEL, TREATED WITH 1.5 PER CENT FERROTITANIUM. B: IRON CONTAINING 0.36 PER CENT CHROMIUM AND 1.2 PER CENT NICKEL, NOT TREATED WITH TITANIUM.

chromium. The features of the molybdenum-treated samples were about as coarse as those of the untreated.

15. The most striking effect of titanium in cast iron as brought out by these tests was the refinement of the graphite particles, or "closing of the grain." In actual practice, the alloy is now being used largely for that purpose.

16. The best results in increasing the strength of cast iron were obtained with titanium and chromium, and it is somewhat surprising to note that the iron alloyed in that way was no more difficult to machine than untreated iron and easier to machine than the more commonly used chrome-nickel or molybdenum irons. This conclusion was checked in different tests, as indicated in Table 1. Furthermore, the iron treated with 1 per cent low-carbon ferrotitanium and 0.5 per cent chromium in the form of the high-carbon alloy is considerably more economical than chrome-nickel or molybdenum-treated iron.

17. These results, therefore, lead to the conclusion that a high-strength cast iron can be made with better machinability and at less cost with titanium and chromium than with chromium and nickel or molybdenum.

DISCUSSION

WRITTEN DISCUSSION

W. H. ROTHER¹ AND V. M. MAZURIE²: It is our belief that the titanium alloy, recently perfected, offers a valuable alloying element to the foundryman for use in cast iron. Due to the graphitizing effect of titanium, the graphite plates formed in a treated iron are very small. This results in a stronger, more uniform, closer grained iron.

The results, shown in Table 3, indicate improvements obtained in a low and a high silicon cupola iron through the use of titanium.

¹ and ² Buffalo Foundry & Machine Co., Buffalo, N. Y.

Table 3
RESULTS OBTAINED FROM TITANIUM ADDITIONS

	Low Silicon Series A		High Silicon Series B	
Ferrotitanium added, %	None	2.00	None	1.00
Silicon, %	0.75	1.10	2.05	2.35
Manganese, %	1.13	1.13	0.64	0.69
Sulphur, %	0.085	0.075	0.145	0.145
Phosphorus, %	0.175	0.180	0.375	0.365
Total carbon, %	3.90	3.90	3.70	3.50
Combined carbon, %	0.68	0.68	0.52	0.48
Graphitic carbon, %	3.22	3.22	3.18	3.02
Transverse strength, lbs.	3250*	4010*	3100	3500
Increase over untreated iron, %	23.4	12.9
Transverse deflection, in.	0.185	0.145	0.180	0.160
Brinell hardness No.	176**	212**	170	183

* Transverse tests were made on 1.25 in. A.S.T.M. arbitration bars broken on a 12 in. span.

** Brinell values show the hardness at a point about 1/16 in. below the skin of the casting.

The metal for all the tests shown in Table 3 was melted in a cupola during a regular daily run. Additions of titanium were made to a small hand ladle with sufficient capacity to pour at least four test bars. For adding titanium to small, hand ladles, we have had most satisfactory results from the nut size material. For larger ladles, the same size works well.

ORAL DISCUSSION

H. BORNSTEIN³: Are the results shown in Tables 1 and 2 individual test results or do they represent average test results?

MR. WELLINGS⁴: An average of six test bars were poured from each mixture.

MEMBER: In what condition is the titanium in cast iron, nitride, carbide or cyanonitride?

MR. WELLINGS: It is probably in both forms.

MEMBER: Have you identified titanium carbide under the microscope? How can you tell the nitride from the cyanonitride?

MR. WELLINGS: We usually find some scattered angular crystals of the carbide. We believe the nitrogen to be present as the cyanonitride.

MEMBER: Do you get any segregation of nitride crystals on the surface?

MR. WELLINGS: We have found invariably that there will be some titanium cyanonitride in all cast irons but we never have found it segregated.

R. S. MACPHERRAN⁵: In the photomicrographs, there are traces of

³ Deere & Co., Moline, Ill.

⁴ Mr. Wellings, Titanium Alloy Mfg. Co., presented paper by request of author.

⁵ Allis-Chalmers Mfg. Co., Milwaukee.

dendritic structure. Does titanium increase the tendency to form dendrites? In some alloy irons there is much dendritic structure.

MR. WELLINGS: I cannot answer your question. That phase of the investigation has not been followed closely.

H. BORNSTEIN: In reference to paragraph 2, stating on the bottom of page 278 and running to page 279, it says "that the chief effect of titanium was to decrease the size of the graphite flakes, and that there also was a tendency to decrease the combined carbon content and increase the formation of graphite and ferrite." I wonder how the author lines up that statement with the statement that it forms a carbide. How can it both form a carbide and act as a graphitizer?

MR. WELLINGS: The titanium is in the form of carbide, but it actually does act as a graphitizer.

MEMBER: Originally this titanium, under the microscope, is there as metallic titanium or titanium nitride, which is more of a base formation than is the carbide. If there is any titanium present as a nitride that will increase the combined carbon.

MR. WELLINGS: We have not noticed the increase of combined carbon.

CHAIRMAN J. T. MACKENZIE*: Titanium belongs to the group of metals, titanium, zirconium and calcium, which have this powerful graphitizing effect, yet, in the case of calcium, it is left only in spectroscopic traces, so there is no indictment of the effect of titanium as to what form the little bit of it left may happen to be.

WRITTEN REPLY BY THE AUTHOR

G. F. COMSTOCK: In reply to the questions regarding the state of combination of the residual titanium found in cast iron, I might explain that we have formed our conclusion that it is present as carbide or cyanonitride from our study of the appearance of these compounds when polished sections are examined under the microscope with white light. The various compounds have been prepared in a reasonably pure form and studied by themselves, as well as mixed intentionally with iron. Pure titanium nitride is yellow, and we have not seen it in commercial iron or steel. Titanium carbide is violet or pale gray, and the cyanonitride, which is merely a carbide containing more or less nitrogen as an impurity, tends to be pink or orange. The nitride and carbide probably are mutually soluble, and may occur mixed in indefinite proportions. The crystals are usually of a very pale color in cast iron, indicating a low content of nitrogen. In steel, they are more reddish, indicating higher nitrogen.

Mr. Bornstein's question as to how titanium can act as a graphitizer and still form a carbide is interesting. The carbide, or cyanonitride as we prefer to call it, is formed at a high temperature, and crystallizes out before the iron freezes, so that it does not unite with iron or form any combination with cementite. The graphitizing effect probably is due to the titanium in actual solution in the iron, which acts partly as a deoxidizer

* American Cast Iron Pipe Co., Birmingham, Ala.

as well as an alloying element. It may be the deoxidation that increases the formation of graphite. At any rate, the graphitizing influence of titanium is an established fact that has been reported by others and has shown up in all our work, in malleable and white iron, as well as in gray iron.

In reply to Mr. MacPherran's question regarding dendritic structures, it is true that the titanium-treated iron often showed this structure in our tests. It was found similarly in other alloy cast irons, especially with chromium. When so much titanium was added as to promote excessive ferrite formation the dendritic structure generally was prominent, but it could also be noticed in the chrome-titanium irons of high strength as is illustrated in Fig. 2.

Gray Iron Production in the Direct-Arc Acid-Lined Electric Furnace

BY CLYDE L. FREAR,* POUGHKEEPSIE, N. Y.

Abstract

This paper describes a method of melting gray cast iron in the direct-arc acid-lined electric furnace. It is the intention of the writer to describe as fully as possible the different operations desirable or necessary to produce soft, easily machinable but close-grained castings, even using a very large amount of steel and cast iron scrap, including chips and borings, and a correspondingly small amount of pig iron in the charge.

1. The use of the electric furnace for melting cast iron has shown a large increase in the last few years. In a number of cases the electric furnace actually is competing with the cupola in melting iron for ordinary castings. The electric process usually is considered more expensive in melting cost per ton than is the cupola process, but for a number of reasons the electrical process may prove to be the more economical in over-all cost.
2. The chief of these economic reasons for preference for the electric process lies in the fact that much cheaper grades of raw material may be used without detracting from the quality of the castings. Many raw materials may be used in the electric furnace that it would be impossible to use in the cupola.
3. A second reason for preference of the electric furnace is the flexibility of the process. The bath of molten metal may be held as long as desired in the furnace and at the same time may be heated to any desired temperature, thus using all the advantages which may result from this high temperature refining and its con-

* Metallurgist, De Laval Separator Co.

NOTE: This paper was presented before one of the sessions on Cast Iron at the 1933 Convention of the American Foundrymen's Association.

trol. In addition, the composition of the bath and the physical properties of the resultant castings may be changed as desired at any time during the melting or the refining process, so that castings of any desired composition or properties may be obtained at will during the process.

4. Furthermore, if a ladle of iron should cool too far, as is often the case when pouring a large number of small castings, it is not necessary to pig this cool metal. It is only necessary to return it to the furnace and reheat it to the desired temperature.

5. A third advantage, especially in periods of low production, lies in the fact that much smaller heats of metal may be made economically in the electric furnace than could be made in the cupola.

6. Another advantage, which is important but which will only be mentioned here, is that with the ease of superheating the metal and consequent high refining and solution of the graphite, ladle additions of alloys should be much more effective than with metal that is only partially refined.

TYPES OF DIRECT-ARC ELECTRIC FURNACES

7. The direct-arc furnace, with which this paper deals, has vertical electrodes extending through the roof of the furnace, the arc playing between each electrode and the bath of metal beneath it. In this type of furnace the heat resulting from the electrical energy is concentrated immediately on the surface of the bath, there being little heating of the bath by reflection of heat from the roof. Every possible precaution is taken, in fact, to prevent the escape of heat from the bath.

8. This type of furnace usually consists of a vertical, cylindrical steel shell with concave bottom. This shell is lined with high-refractory material of desired composition and of sufficient thickness to prevent the heat within from penetrating so as to heat the steel shell to a temperature high enough to melt or otherwise destroy it. The roof is built up from special silica shapes with the required openings for the electrodes. The furnace shell and top are arranged for water cooling at all openings or in any place where overheating is liable to occur.

9. Most furnaces are of the three-phase type, requiring three electrodes, although some smaller furnaces and some of the older ones are of single-phase type and have only two electrodes. The electrodes are so arranged that the distance of the lower ends from the charge is controlled by the current flowing through the

cables leading to them, thus allowing an almost constant current input at all times. The control board is equipped with an automatic cut-out to prevent excessive current input, and also with regulators to control the input at all times.

10. In addition, there is an apparatus to control the speed of raising and lowering the electrodes as well as allowing the electrodes to be raised or lowered by hand or thrown into automatic control as desired. A high and low tap on the transformer, with a switch on the control board, allows the use of full or lowered voltage as desired during the melting.

11. The furnace lining itself may be acid, basic, or neutral. The acid-lined furnace, which is generally employed in the gray cast iron industry, has a bottom built up of rammed siliceous material, usually crushed ganister. The side walls and roof are built up with high-refractory silica shapes. During operation with an acid hearth there is always a tendency toward reduction of the silica of the hearth by the carbon of the metal, probably according to the equation



12. There is always present, therefore, a tendency toward an increase in the silicon content of the metal and a corresponding decrease in the carbon. However, if the bottom is correctly built, solidly rammed, and kept free from holes and loose pieces, this action should be very slight.

13. The basic lining is made by building up the bottom with burned magnesia or dolomite. The walls are lined with magnesia brick to a point above the metal and for the remaining distance, including the roof, with the acid silica brick. A row of chrome or other neutral brick must be laid between the acid and the basic brick to prevent interaction. In the basic process, unless great care is taken to exclude all rusty scrap, and to keep the furnace atmosphere reducing at all times, there will be a tendency to reduce the silicon and perhaps the carbon. A neutral furnace would be lined with chrome or alumina brick to a point above the surface of the bath. As far as the author knows, there is no production installation using a neutral lining.

14. It will be the purpose of this paper to cover only the melting of unalloyed gray cast iron in the acid-lined direct-arc furnace, using raw material charges containing as little pig iron as possible, and to explain the reason for each step as fully as a paper of this length will allow.

METALLURGICAL PROBLEMS INVOLVED

15. When considering the metallurgical problems involved in the electric melting of gray cast iron, it must be remembered that the properties of the finished casting are almost entirely dependent upon the amount, distribution, shape and size of the graphite particles. In other words, we may have several gray cast irons of identical chemical composition, even to the content of combined carbon and graphite, yet these irons will have different physical properties and machining qualities. The chief items of control, therefore, in the production of iron with specified physical properties are first, the composition of the bath itself and, second, the control of the graphite precipitation in the casting when poured.

16. The composition of the bath is controlled by the composition of the original charge and subsequent sub-charges, and by chemical reactions taking place between the bath and the bottom, the bath and the slag, and in the bath itself. When steel or low-carbon pig iron or scrap is used in the charge, the deficiency in carbon must be made up by the use of some material which is high in carbon, such as petroleum coke. For rapid solution of this carbon, the coke should be placed as near the bottom of the charge as possible, with all the steel above it. When charged in this way, all melted metal must flow over the coke, dissolving the maximum amount of carbon.

17. As silicon reduces the solubility of carbon in iron, no ferrosilicon should be added with the original charge. When the bath is melted, the coke will rise to the surface but carbon absorption will continue at a reduced rate. The time required for the absorption of the carbon can be determined only by experience, but will increase with rise in temperature of the metal.

18. With higher bath temperatures, however, there will be an increased tendency toward reaction of the metal with the acid bottom, thus increasing the silicon and tending to reduce the rate of carbon solution. It will be evident, then, that the temperature of the bath should be kept as low as possible, but high enough to cause the absorption of carbon at the desired rate.

19. As has been already mentioned, if the bottom is correctly built and properly burned-in, this silicon pick-up will be low and can be easily kept under control. In long heats, as in continuous melting, there will always be a slight continuous silicon pick-up due to interaction with the acid bottom, and this must be counteracted by the addition of steel which, in turn, requires additional

coke to hold the desired carbon content. The rate of silica reduction and the necessary amount of steel and coke to be added must be determined by experience, but when once determined, the composition of the bath can be held constant as long as desired.

GRAPHITE CONTROL

20. With few exceptions, it may be considered that the properties of gray cast iron depend almost entirely upon the control of the graphite. Cast iron may be likened to a medium carbon steel containing a very large number of plate-like inclusions, these inclusions being made up almost entirely of graphite. The strength of the iron will depend to a certain extent upon the various constituents of the metallic matrix, just as the strength of steel varies with the carbon content, the grain size and the form of the various constituents. The effect of these latter variables, however, is much less than the effect of the size, form and number of the graphite inclusions.

21. These graphite inclusions, because of the absence of any strength of their own, weaken the metal due to breaking up of the continuity of the metallic matrix, and also weaken it under impact loading due to the notch effect at the edge of each graphite flake. This same notch effect is the cause of the brittleness or, more correctly speaking, the lack of ductility in the metal. If the graphite exists only as rounded particles or nodules, the weakening due to discontinuities of the metallic matrix and the brittleness due to the notch effect will be much less for the same amount of graphite, and hence the strength under both static and dynamic loading will be much increased. We may have almost any size and shape of graphite inclusion, from the very large, overlapping plates present in very soft and porous iron to those small nodular particles present in high-grade malleable cast iron.

22. When a soft pig iron containing very large graphite flakes is melted in a hearth-type furnace such as the air furnace, the open-hearth furnace or the electric furnace, the graphite immediately starts to dissolve in the molten iron, the speed of solution increasing with the rise in the temperature of the bath. The amount of graphite solution also is proportional to the time of holding and inversely proportional to the amount of carbon already dissolved in the iron. The rate of carbon solution will therefore decrease as the amount of dissolved carbon increases unless the temperature of the bath is raised or the iron stirred. If sufficient time is allowed, however, the graphite will be entirely dissolved,

unless, of course, the metal becomes saturated with carbon before this solution is complete.

23. The graphite distribution in the solid iron will depend to a large extent on the number and size of the undissolved graphite particles in the molten iron before pouring. If the soft pig iron mentioned above is simply melted and immediately poured, most of the graphite originally present in the pig iron will be present in the molten iron. The large flakes originally present will have become broken up in the melting process, and as there has been little time allowed for the solution of this graphite, the number of flakes will have materially increased. Each of these graphite particles present in the molten iron will act as a nucleus for the start of a new flake which will form in the cooling and solidifying iron.

24. If this iron cools slowly, as in sand molds and in heavy sections, the graphite flakes will be large and overlapping and the casting will be very soft and probably porous. If the iron is cooled quickly, the graphite flakes necessarily will be much shorter and the casting considerably harder and closer grained.

25. If, however, the melt of the above soft pig iron is held for a considerable time at a comparatively high temperature, the graphite originally present will be entirely dissolved, and as a result, there will be no graphite nuclei in the metal when poured. As a certain number of these nuclei are necessary to cause precipitation of graphite from liquid metal and probably from metal in the pasty state, a metal containing no graphite nuclei probably will cool to a temperature considerably below that of solidification before any graphite separation occurs. When the temperature reaches that of a dull red heat, graphite separation will suddenly occur at a very rapid rate and with evolution of considerable heat.

26. This graphite which separates from solid iron will be in the form of very fine short plates or in the form of nodules which are really bundles of such fine plates. As these fine plates and nodules are much less effective in breaking up the continuity of the metallic matrix, such iron will be much stronger and more ductile than that melted without refining. Between these two extremes we may produce any grade of iron by controlling the temperature of the bath and the holding time, thus controlling the graphite distribution in the molten iron.

27. When steel is used in the charge, this will, of course, add no graphite to the melt. Necessary additions of carbon are made by the addition and solution of petroleum coke. This, having no crystal form of its own, will not form nuclei to aid in the graphite

separation. The purpose of the coke is only to make up the required total carbon content in the bath. As a result, the addition of steel to the bath, even when the carbon content is increased by the use of coke, will serve only as an additional refining and the result will be to produce an iron of higher strength. Such addition of steel, with the addition of the necessary coke and ferro-alloys, serves as a convenient way of quickly refining the metal in the bath and is, in fact, often used where it is necessary to reduce quickly the number of graphite nuclei in the molten iron.

28. Long holding, especially at higher temperatures, often will cause excessive refining of the iron. While this is desirable, or perhaps necessary when making high-strength castings, or with heavy castings, it is not desirable where high strength is not a necessary consideration or where the castings must be thin but easily machinable. A further result of such refining is that it may cause excessive shrinkage in castings where allowance for such shrinkage has not been made in molding.

29. To correct such over-refining, it is necessary to increase the concentration of graphite nuclei in the molten metal, without, however, changing the composition of the bath. This is most effectively accomplished by the addition of soft pig iron or soft cast iron scrap the composition of which is the same as that of the bath, in sufficient quantity to correct the shrinkage or to effect the necessary softening. After such addition the heating should be only sufficient to bring the temperature of the iron to that required for pouring. Such heating should be done under low power to avoid undue refinement immediately under the electrodes.

30. At this point it would be well to mention the fact that thorough mixing of the molten charge is effected in this type of furnace without rocking or rotating the furnace itself. As has already been mentioned, the heat is applied immediately on the surface of the metal under the lower end of the electrodes by the arc playing between the electrodes and the bath. The electric current, therefore, must flow through the metal and induce magnetic forces therein. These magnetic forces set up convection currents in the metal bath, causing thorough mixing without contamination by slag or by material from the walls. Because of this heating immediately under the electrodes and the exchange of this metal for cooler metal, refining of the whole bath is accomplished at a low temperature, which would take a much longer period or a much higher temperature with any other method of applying the heat.

RAW MATERIALS

31. The raw materials usually available and which may be used in the direct-arc furnace may be tabulated as follows:

- (A) Pig Iron.
- (B) Cast Iron Scrap.
 - (1) Returns—
 - (a) Heavy pouring gates, heavy castings, pigged metal.
 - (b) Light.
 - (2) Automotive Scrap.
 - (3) Machinery Scrap—
 - (a) Heavy.
 - (b) Light.
 - (4) Miscellaneous.
 - (5) Borings.
- (C) Steel Scrap.
 - (1) Bar Stock.
 - (2) Sheet Cuttings.
 - (3) Miscellaneous.
 - (4) Turnings.
- (D) Alloys.
 - (1) Ferrosilicon.
 - (2) Ferromanganese.
 - (3) Ferrophosphorus.
 - (4) Ferrochrome.
 - (5) Nickel.
- (E) Coke.
- (F) Fluxing Materials.
 - (1) Lime.
 - (2) Fluorspar.

Pig Iron.

32. In making up a furnace charge, regardless of the type of melting equipment used, it is desirable to use as little pig iron as possible and still obtain the desired properties in the castings. Due to its lower cost, it is desirable to use as much scrap as possible in the charge. As it is possible to make excellent cast iron without using any pig in the electric furnace charge, it is not necessary to stock a large variety of pig irons. The chief use of pig iron is to add graphite nuclei. It should not be necessary, therefore, to stock more than two grades of pig iron, both of them soft, having the approximate composition of the iron which is to be produced. Where possible, it is best to have the silicon contents of these irons as low as possible consistent with high graphite content.

Cast Iron Scrap.

33. The foundry returns should be segregated so as to keep heavy material separate from the light pieces. Heavy returns, such as heavy pouring gates, heavy castings and any metal which has been previously pigged (especially if made from iron melted for light work), should be soft and contain a large percentage of graphite. For this reason, and because it is of the same composition as that usually melted in the furnace, it will serve admirably for softening the charge. Light returns, preferably, should be used in the original charge or in early subcharges.

34. All returns, including gates, risers and pigs should be tumbled to remove the adhering sand. Any sand adhering to the metal will rise to the surface of the bath when the metal is melted, and in this position will not only be subjected to the intense heat of the arc but will be in contact with the coke floating on the bath. Conditions will be perfect, therefore, for the reduction of this silica to silicon, which will immediately dissolve in the molten metal. While this probably would not be so harmful in the original charge, provided all sand or silicates were raked from the surface of the metal as soon as it was melted, it would make control of the silicon very difficult, if not impossible, if added with subsequent subcharges.

35. Nothing need be said about automobile scrap except that it is usually of excellent grade and of fairly constant composition, and makes a good material to replace pig iron.

36. Machinery scrap makes an excellent raw material for electric melting, provided the phosphorus and the sulphur are within the allowable limits. A large proportion of the machinery scrap is of large cross-section and contains a large percentage of graphite, and will serve as an excellent graphitizer and shrinkage preventor if its composition is approximately the same as that of the bath.

37. Many papers have been written on methods of using cast-iron borings. No difficulty has been encountered by the writer in using this material in the electric furnace. At our plant the foundry is remelting all cast-iron borings resulting from the manufacturing operations on the total output of the foundry. No exact data have been obtained on the melting loss when borings are used, but this loss does not exceed ten per cent. The borings are not subjected to any treatment before melting but are used just as they come from the shop.

Steel Scrap.

38. It is obvious that the sulphur content of this class of material should not be too high. Bessemer screw stock and free-cutting steels have a sulphur content considerably higher than that desired in cast iron, and when these are used care must be taken not to add them in sufficient quantity to increase this element to a point above the allowable maximum. Of course, all alloy steel scrap should be segregated for use in making alloyed cast iron.

39. Steel chips, as they come from manufacturing operations, usually are bulky and covered with cutting lubricant. Their bulkiness will make them very inconvenient when charging, and the oil will cause an excessive amount of smoke. The company with which the writer is connected treats the chips as soon as they come from the machine shop by first crushing them and then removing the oil by centrifuging. It might be added that this method was used long before the foundry was built, as the oil which was reclaimed and the saving in shipping costs of the crushed chips more than repaid the cost of the reclaiming and crushing operations.

Alloys.

40. Ferrosilicon, ferromanganese and ferrophosphorus usually are added in the furnace and should be crushed to about the size of walnuts or slightly finer. Ferrochrome and nickel, which are usually added in the ladle, should be crushed to 10-mesh or finer. While the latter two alloys are considered as alloying elements, they are often useful when making straight carbon castings where it is desired to pour a small number of castings with heavy sections without having to resort to a drastic lowering of the silicon content of the entire charge in the furnace.

Coke.

41. Petroleum coke is used in electric melting almost exclusively as a carburizer. This coke is formed during the process of cracking petroleum, and since it is produced from gaseous material it might be considered to be almost colloidal as compared to ordinary coke, which retains to a certain degree the crystal form of the coal from which it was formed. In addition to its high solubility in the molten iron, its sulphur content is very low. This material should be crushed to about walnut size.

Fluxing Materials.

42. For fluxing and building a slag on the bath, powdered quicklime and crushed fluorspar should be used. Sand, al-

though very efficient in imparting fluidity to the lime slag, should never be used for this purpose due to the difficulty of controlling the silicon content in the metal when this material is used. This is explained above under the use of foundry returns.

PREPARATION OF THE ACID-LINED FURNACE

43. The direct-arc furnace is so built that the electrodes can be removed and the top lifted off. In building the new lining or when relining the furnace, crushed ganister or other suitable siliceous material is moistened with an organic binder such as molasses, and this material is shoveled onto the bottom. It is then solidly rammed in place, preferably with an air rammer having a head similar to a sand rammer. When an air rammer is not available a hand rammer may be used, but the operation will take longer and it will be more difficult to exert the required pressure.

44. A small amount of material should be added at a time and each addition rammed as solidly as possible. This addition and ramming is continued until the thickness of the bottom is about eighteen inches, and when the edges are leveled off these edges should be almost even with the bottom of the pouring spout and the bottom of the charging door.

45. When the bottom is completed the side-walls are built up of silica brick so tapered that, when placed closely together, they will form a circle the outside diameter of which is about two inches less than the diameter of the furnace shell. No mortar is used between the bricks, which are laid in close contact with each other. When a course is laid, fine silica sand is brushed over the tops of the brick in such a manner as to fill all spaces between them. The space between the bricks and the furnace shell is filled with sand or loosely rammed ganister. The furnace wall is completed in this manner, using special shapes for pouring spout and door arch. The roof then is set in place and electrodes connected.

46. When the furnace is completed it is necessary to burn in the bottom. To do this a layer of coke is placed on the bottom, or two pieces of broken electrodes are so placed in the form of a cross on the bottom that the three electrodes will touch them, thus making electrical contact with each other. The current is then turned on and the electrodes lowered until an arc is struck between them and the electrodes or coke on the bottom. Automatic electrode control is used and the current input kept as low as possible until the bottom is thoroughly dried.

47. The current is then increased sufficiently to cause the

bottom to sinter into a solid mass to a depth of two to three inches without causing excessive melting of the brick walls and roof. There will be a slight softening of the walls and roof so that finally there will form a number of small stalactites, especially just over the charging door. The formation of these stalactites is an indication that the burning operation has been completed, provided the current input has been kept as low as possible during the operation.

48. When the bottom has been built in this way there should be no trouble due to pieces breaking off and rising to the surface of the molten metal, or due to excessive reaction between the bottom and the bath. In fact, the bottom should last for a large number of heats before patching is required.

49. When a large number of heats have been melted it may be noted that holes are being formed in the bottom or that small pieces loosen and rise to the surface of the metal. When this occurs, it is necessary to resort to patching, which is done by adding more bottom material and burning it into place. When the furnace is emptied and still hot, crushed ganister (which may be mixed with a small amount of fireclay) is thrown in with as much force as possible, filling all holes and producing a perfectly smooth saucer-shaped hearth. After the addition of the patching material the bottom is again burned in as described in paragraph 46, until all new material is solidly sintered into place and any cracks are sintered together.

MELTING

50. The following description of the melting process will be based on the assumption that little or no pig iron is to be used except for softening or preventing shrinkage. The original charge will be composed, therefore, of returns, cast iron scrap, steel scrap including chips, and cast-iron borings.

Charging.

51. When the furnace is cold the coke should be charged directly on the bottom. The amount of coke required for chemical combination in the iron should be calculated and at least 25 per cent more than this amount added. The cast iron scrap, including the returns, should then be added, and on top of this the steel. The steel chips should then be shoveled in, spreading them as much as possible over the previously charged metal. The cast-iron borings should be added last, as these will serve to make a better electrical contact with the electrodes.

52. The purpose of this method of charging is to place the carbon at as low a point as possible in the charge so that the molten iron, as it trickles to the bottom, will have a chance to absorb the maximum amount of carbon. For this same reason no silicon should be added at this time. When charging a hot furnace it will be impossible to add the coke first. The order of charging in this case would be: cast iron scrap and returns, coke, steel scrap, steel chips, and cast-iron borings.

Melting Down the Charge.

53. After charging is completed the current may be turned on with a low tap and the input cut to as low a point as possible and still produce a nearly constant arc with as little surging as possible. At first it may be necessary to poke the charge with a slice bar to hold the arc. If the current input is too high at first, the electrodes will bore their way through the charge and may sinter it together sufficiently to almost prevent readjustment of the material in the furnace to correct this condition. If the electrodes persist in boring into the charge, this may be prevented by placing some heavy steel scrap across the top of the charge to form a conductor until the top of the charge melts sufficiently to form its own conductor and to prevent boring and surging.

Control of Slag.

54. The heating should be continued with low current input until the whole charge is melted. It may be necessary during the melt-down to push unmelted pieces of metal from the edges to the center of the bath. After melting is completed the electric current input should be increased and the temperature of the bath brought to about 2600 degs. Fahr. At this point all excess coke and slag which has formed on top of the bath should be raked off and sufficient lime added to make a very thin slag blanket.

55. A small amount of fine petroleum coke also should be added, which will unite slowly with the lime to form a slightly carburizing slag. The purpose of this slag is, of course, to aid in purifying the metal, to effect slight carburization and to prevent oxidation, but it also prevents excessive reflection of heat from the surface of the metal. Such reflection not only would cause excessive softening of the roof but would prevent rapid heating of the metal.

56. The slag blanket, therefore, should cover the entire bath. If it is allowed to become gummy it will be carried to the edge of

the bath, leaving the metal in the center exposed. To prevent this, sufficient fluorspar should be added to impart the necessary fluidity. As has already been mentioned, sand should never be used for this purpose.

Finish Melting.

57. As soon as the slag has been formed and before the temperature of the bath has been allowed to increase, a sample bar should be poured for analysis. A convenient bar measures $1 \times 1 \times 5$ inches and is poured vertically in an open sand mold. This bar is allowed to solidify in the sand and is then removed and allowed to cool in air until the color is a dull red or until its temperature has reached a point below the hardening range. It is then quenched in water, drillings taken, and these analyzed for total carbon and silicon and, if necessary, for manganese and phosphorus.

58. The carbon and silicon determinations can be made by one man within forty minutes from the time the test bar is first poured. With a well-made bottom and with no silica in the slag, there should be no change in the chemical composition of the bath if it is held at a low temperature during this period (see Table 5). There will be, of course, a certain amount of graphite refinement, but if the temperature is kept low with only sufficient power input to hold this low temperature, the graphite refinement will produce but slight change in the properties of the iron when cast.

Adjusting the Composition.

59. With a small amount of experience it should be possible for the melter to obtain almost exactly the composition desired in the first melt-down, regardless of the composition of the original charge. In other words, it should be possible to use almost any reasonable combination of raw materials and, by proper manipulation during melting, produce the desired composition in the bath when melted. It may be necessary to add ferrosilicon and ferromanganese, but it should not be necessary to adjust the carbon to any great extent.

60. Any considerable adjustment of the carbon is undesirable, especially when making soft castings. If the carbon content is high it must be reduced by adding steel to the bath, thus reducing the graphite concentration by dilution. If, on the other hand, the carbon content is low, more coke must be added and the heat increased to effect the necessary solution of carbon. This high

heat causes further refinement by solution of the graphite nuclei. It is desirable, therefore, to manipulate the melting so that the carbon content will be that desired in the molten bath, and adjust the content of the other elements.

61. If it has been necessary, while correcting the composition, to refine the iron more than is desired, this excess refinement may be corrected by the addition of pig iron, heavy returns or heavy scrap, the composition of which is approximately the same as that of the bath. This addition of soft iron will supply the graphite nuclei which have been removed by refining.

Raising to Finish Temperature.

62. When the bath has been corrected to the desired composition it is heated as rapidly as possible to the pouring temperature, without, however, using too high a current input. When the current input is too high the metal immediately under the electrodes will be over-refined. As soon as the bath reaches the correct temperature for pouring, it is well to cast a wedge-shaped specimen or stepped casting. This, when broken at the desired cross-section, will indicate the amount of refining that the bath has undergone.

63. Judging from the preliminary analysis and the fracture, the necessary amount of conditioning can be rather accurately estimated. If it is desired to make high-strength castings and the test bar shows a grayish-black fracture at the desired cross-section, it will be necessary to further refine the bath to effect further solution of the graphite. This may be accomplished by holding the metal in the furnace at approximately 2700 degs. Fahr. until the test bar shows the required fine-grained silvery fracture.

64. If, on the other hand, it is desired to obtain soft, easily machined castings, and the test bar shows a very fine-grained silvery fracture, indicating excessive refinement, it will be necessary to increase the graphite concentration as described above, by the addition of soft pig or scrap. It is evident that if the iron used has the same composition as the bath no change will be made except the required increase in the concentration of graphite nuclei.

Continuous Melting.

65. When making continuous melts in the electric furnace, the bath is first built up to the desired amount as described above; then, when a ladle of metal is removed from the furnace, a cor-

DIRECT-ARC ACID ELECTRIC FURNACE FOR GRAY IRON

Table 1
FURNACE RECORD AND COMPOSITION OF ELECTRIC FURNACE HEAT

Charges		Material Charged, Lbs.				Chemical Composition					
Charge No.	Time.	Returns.	Steel Scrap.	No. 1 Machinery Scraps.	Auto-mo-bile Scrap.	Pig Iron.	Fe-Si (50%)	Fe-Mn (80%)	Fe-P (23%)	Coke.	
Orig.	9:45	484	836	1100	1023					66	
1	12:30	627					31.8				
2	2:40	264	88								
3	3:46	385									

* Test bar "D" cooled overnight in sand.
Poured on low tap at 11:40 A. M.
Metal poured at 2:00 P. M. and 4:00 P. M.

Test.	Time.	Total Carbon.	Graphite.	Combined Carbon.	Silicon.	Manganese.	Phosphorus.	Sulphur.
A	1:00	3.38	***	***	2.27	0.50	0.42	***
B	2:45	3.32	***	***	2.20	***	***	***
C	3:00	3.35	***	***	2.20	***	***	***
D*	4:30	3.40	2.83	0.57	2.24	0.67	0.51	***

Table 2
FURNACE RECORD AND COMPOSITION OF ELECTRIC FURNACE HEAT

Charge No. Orig.	Time.	Returns.	Steel Scrap.	Cast Iron Botings.	Material Charged, Lbs.		Fe-Mn (80%)	Fe-Si (50%)	Fe-P (23%)	Coke.
					No. 1 Miscel. laneous Scrap.	Pig Iron.				
1	8:00	1430	616	44	362	792	88	13.2	66	11
2	12:20	264	88	44	88	88	8.8	11.0	11	11
3	2:30	264	88	44	88	88	9.9	11.0	11	11

CHEMICAL COMPOSITION										
Test.	Time.	Total Carbon.	Graphite.	Combined Carbon.	Manga- nese.	Silicon.	Phosphorus.	Sulphur.		
A		3.39	0.51	0.44		
B		3.22	2.10		
C		3.57	2.24		
D		3.40	2.32		
E*		3.33	2.72	0.61	2.31	0.65	0.50		

* Test Bar "E" cooled overnight in sand.
Power on low tap at 8:30 A. M.
Metal poured at 11:00 A. M. and 4:00 P. M.

responding amount of cold material is added to the bath. In making these subcharges, care must be exercised to so proportion the raw materials as not to upset the balance of carbon, silicon, graphite nuclei and other alloying elements.

66. There naturally will be a tendency to an increase in the silicon content of the bath, and this must be corrected by the necessary addition of steel. This steel, in turn, will require a higher carburizing action, which can be obtained only by increasing the amount of coke in the slag and increasing the temperature, especially under the electrodes. This increase in temperature will necessarily cause excessive refinement of the graphite in the bath, and it will be necessary to correct this by the addition of pig iron or heavy scrap. It might be stated that the greatest difficulty in continuous melting results not from a change in the composition of the bath but from failure to keep a sufficient concentration of graphite nuclei in the bath.

Typical Melts.

67. Tables 1 and 2 are made from actual melting charts for two days, a slightly different melting process being carried out each day.

68. Table 1 represents a heat of much shorter duration than that represented by Table 2. It will be noted that no pig iron was used in the original charge and that no silicon was added until the original charge was melted and the desired amount of carbon dissolved. At 2:40 P. M. a small amount of steel scrap was added, along with the returns to reduce the silicon content, and some pig iron to counteract the refining due to holding two hours and to the addition of the steel. It will be noted that the total carbon and the silicon was held within very close limits during the entire heat, and that the combined carbon is normal for a high grade, soft but close-grained iron. This iron was melted for castings which must withstand considerable hydraulic pressure and yet must be easily machined.

69. In the heat represented by Table 2 a smaller total amount of iron was charged but the melting period extended over a much greater length of time. After a heat was poured off at 11:00 A. M., the metal remaining in the furnace was carried until 4:00 P. M. before pouring. During this period steel was added to reduce any silicon, and two charges of coke were thrown in and the heat increased so that there was an increase of 18 points in the total carbon. This was reduced with steel and the silicon increased to

Table 3
DATA ON HEAT MADE IN 1½-TON PER HOUR FURNACE

	Lbs.	Per Cent of Metallic Charge
Original Charge.		
Foundry Returns.....	704	17.2
Steel Scrap.....	816	20.0
No. 1 Machinery Scrap.....	1320	32.3
Automobile Scrap.....	1243	30.5
Petroleum Coke.....	77

Table 4
ADDITIONS MADE TO BATH TO CONDITION FOR FLOOR POURING

Additions.	Time Made	
	3:20 P. M.	3:45 P. M.
Foundry Returns.....	264 lbs.	264 lbs.
Steel Scrap.....	88 "	88 "
Cast Iron Borings.....	44 "	44 "
Pig Iron.....	88 "	88 "
Ferrosilicon (50%)	27.7 "	15.4 "
Ferromanganese (80%).....	3.7 "	4.0 "
Ferrophosphorus (23%)	18.5 "	18.5 "

the desired point with ferrosilicon. Considerable pig iron was added to counteract the long refining. The composition of the pig iron used was as follows:

Total Carbon	3.75%
Silicon	2.42%
Manganese	1.01%
Phosphorus	0.64%
Sulphur	0.044%

Effect of Holding at Low Temperature.

70. To show the effect of holding at low temperature in the furnace, a complete description of one heat made in a 1½-ton per hour furnace is given in Tables 3, 4 and 5. The current was turned on using the low tap at 11:00 A. M. and continued with low tap until 1:00 P. M., when the bath was completely melted. At this time all slag was raked from the surface and a thin lime slag added. The power was kept on low tap and the current input cut to the minimum necessary to hold the arc. Standard A.S.T.M. 1.20 × 21 inch test bars were poured at 1:30 P. M. These bars were cast in green-sand molds placed with the riser end about two inches above the pouring end.

71. Further test bars were poured after the bath had been

Table 5
EFFECT OF LOW-TEMPERATURE REFINING ON THE COMPOSITION AND PHYSICAL PROPERTIES OF ELECTRIC CAST IRON

Test	A	B	C	D	E	4:00
Time, P. M.	1:30	2:00	2:30	3:15	3:40	
Time of Refining, Hours	None	1/6	1	1 1/4	(Note 1)	
Temperature, degs. Fahr.	2500	2530	2570	2700	2650	
CHEMICAL COMPOSITION						
Total Carbon, %	3.57	3.61	3.63	3.64	3.57	3.63
Graphite, %	2.86	2.83	2.87	2.89	2.83	3.05
Combined Carbon, %	0.71	0.78	0.76	0.75	0.74	0.58
Silicon, %	2.23	2.20	2.20	2.20	2.26	2.33
Manganese, %	0.58	0.66	0.56	0.54	0.69	0.70
Phosphorus, %	0.38	0.37	0.37	0.38	0.49	0.48
Sulphur, %	0.067	0.061	0.064	0.064	0.062
PHYSICAL PROPERTIES						
Tensile Strength (lbs. per sq. in.)	18,280	20,240	20,510	22,000	22,660	
Compressive Strength (lbs. per sq. in.)	69,080	72,610	72,170	85,330	84,990	
Briell Hardness	131	134	143	149	149	149
Drill Test, Keep's (rev. per inch)	388	431	425	413	428	
Machinability (inches penetration per 100,000 revolutions)	258	232	235	242	234	

NOTES: (1) Test taken after first sub-charge. (See Table 4.)
 (2) Test taken after second sub-charge (1 x 1 x 5 in. bar). (See Table 4.)

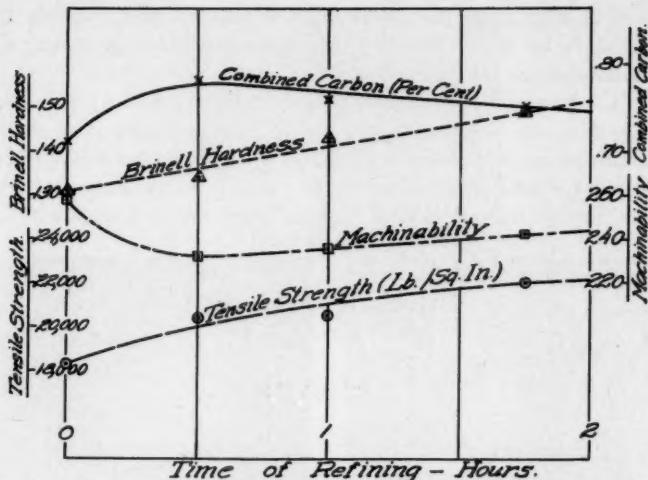


FIG. 1—EFFECT OF LOW-TEMPERATURE REFINING ON THE PHYSICAL PROPERTIES OF ELECTRIC-FURNACE CAST IRON.

held at low temperature for $\frac{1}{2}$ hour, 1 hour, and $1\frac{3}{4}$ hours. A fifth set of test bars was cast at 3:40 P. M. after the bath had been conditioned for floor pouring by the first of the additions shown in Table 4. A one-inch square test bar was cast fifteen minutes after the last addition for analysis only.

72. From Table 5 it will be seen that almost no change took place in the composition of the bath during the $1\frac{3}{4}$ -hour holding. Even the combined carbon content can be considered very nearly constant within the allowable limits of chemical analysis (specimens *A* to *D*, inclusive).

73. There will be noted a distinct increase in the tensile strength and the Brinell hardness. This increase in tensile strength and hardness does not coincide, however, with the change in machinability¹ as determined by the drill test. It is usually considered that an increase in the tensile strength and hardness causes a decrease in the machinability, but as shown in Table 5 and the

¹ In Table 5 are given two machinability readings. The drill-test figures are obtained with the regulation Kepp hardness tester using a 3/8-inch straight fluted drill under a load of 150 lbs. The number given represents the revolutions per inch penetration of the drill. The same drill was used without resharpening between tests. A new drill was first used under test conditions on a piece of soft cast iron, then two drill tests made on each test piece. If the two curves obtained were not parallel in all tests, a new drill was obtained and the whole series of tests repeated.

The machinability value was obtained by dividing 100,000 by the drill machinability (revolutions per inch penetration). The advantage of this value lies in the fact that improved machinability is expressed by a higher instead of a lower number.

curves in Fig. 1, the machinability decreased slightly during the first half hour of refining and then increased steadily during the remainder of the refining period.

74. A study of the microstructure (Figs. 2 and 4) will show one reason for the increase in strength. A comparison of the graphite in the iron when first melted with that after refining for 1½ hours shows a large reduction in size of primary graphite flakes due to refining. Either there were fewer graphite flakes

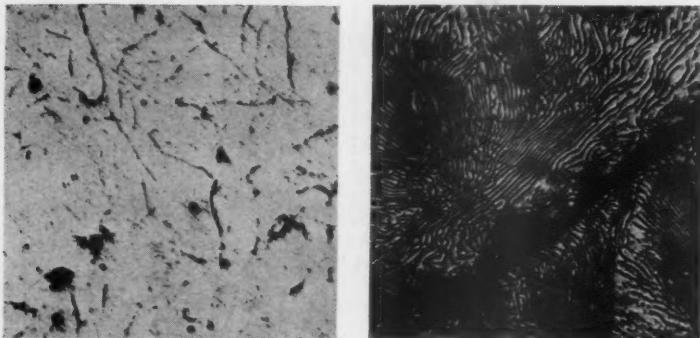


FIG. 2 (LEFT)—MICROSTRUCTURE OF ELECTRIC-FURNACE GRAY CAST IRON. SAMPLE TAKEN PREVIOUS TO REFINING. X100, UNETCHED, SHOWING PRIMARY GRAPHITE.

FIG. 3 (RIGHT)—SAME SAMPLE AS FIG. 2. X1000, ETCHED IN 2% HNO₃, SHOWING PEARLITE AND PRIMARY GRAPHITE.

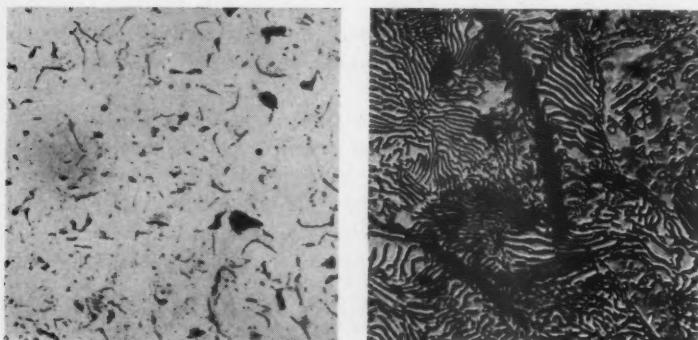


FIG. 4 (LEFT)—SAMPLE OF ELECTRIC-FURNACE CAST IRON, TAKEN AFTER 1½ HOURS REFINING. X100, UNETCHED, SHOWING PRIMARY AND SECONDARY GRAPHITE.

FIG. 5 (RIGHT)—SAMPLE OF ELECTRIC-FURNACE IRON, AFTER REFINING 1½ HOURS. X1000, ETCHED IN 2% HNO₃, SHOWING PEARLITE, PHOSPHIDE AND PRIMARY GRAPHITE.



FIG. 6—SAMPLE OF ELECTRIC-FURNACE IRON TAKEN AFTER 1½ HOURS REFINING. X1000, ETCHED IN 2% HNO₃, SHOWING PEARLITE, FERRITE, DECOMPOSING PEARLITE GRAINS, AND BOTH PRIMARY AND SECONDARY GRAPHITE.

formed in the molten and pasty iron or there was less chance for growth of these flakes before the iron solidified. It probably is safe to assume that both conditions occurred as a result of the solution of a large proportion of the graphite nuclei present in the iron when first melted.

75. In the refined iron, however, there are numerous secondary graphite particles, as shown by the small rounded black areas in Fig. 4, and which appear as comparatively thick angular particles under higher magnification (Fig. 6). Further proof of this secondary separation of graphite particles lies in the fact that each of these secondary graphite particles is in immediate contact with ferrite, or grains of pearlite which have been almost completely decomposed.

SUMMARY

76. When all operations are conducted correctly, a soft, easily machinable but close-grained iron may be economically melted in the acid-lined direct-arc electric furnace. The main points to be stressed are as follows:

- (A) Correct building up and ramming of the hearth, and the repairing of the same.
- (B) Elimination of silica from the slag on the surface of the bath.
- (C) Melting in such a manner as to have the carbon content of the molten bath very close to the desired value, therefore requiring very little reconditioning of this constituent.

(D) Control of the graphite nuclei in the molten bath, to produce the strength, machinability and other properties desired.

77. The number and size of the graphite nuclei are dependent upon the graphite concentration of original charge and upon the temperature of the bath and the time of holding in the molten condition. It is by control of these variables that the final properties of the castings may be predetermined.

ORAL DISCUSSION

H. S. AUSTIN:² We are using electric furnaces and are having very good results, especially on our centrifugal brake drums, where we are using the duplexing process to get a uniform chill. We take a chemical analysis and make chill tests every few minutes, holding the results practically uniform throughout the heat. The furnace operates an average of about 18 hours per day. On Fridays and Saturdays, it sometimes is operated as high as 31 continuous hours. The furnace is kept filled to full depth all the time and we do not have any time for repairs during a run. Repairs are made during the week ends.

MEMBER: We have a small 100-kilowatt, electric furnace and operate it 24 hours a day, every day. The molders work right along with the furnace. We operate three crews, 8 hours a day. The melting cost on this furnace has been reduced to about \$7.00 per ton. We have found this electric furnace to be about as cheap as the cupola. This is in the West where the price of coke is higher than in the East. Iron costs are about the same as in the East. Scrap costs us approximately \$5.00 per net ton. We have a power charge on the 100-kilowatt furnace of \$112 per month regardless of the amount of power we use.

J. T. MACKENZIE:² I have not had much experience with electric furnaces lately. Our principal trouble in running special alloys in the furnace, is hitting the carbon and silicon contents. Speaking of the direct-arc furnace, we find that with the graphite electrode we must have one time schedule and with the carbon electrode another. If the mixture is set to produce 2.80 per cent carbon and 2.50 per cent silicon, and there is some hold-up which requires that the current be kept on for another 20 minutes, it is necessary to raise the carbon and lower the silicon by a low silicon pig iron addition.

We have had to set up a rather elaborate schedule for the different alloys to be able to secure the analysis we desire at a certain temperature.

² Campbell, Wyant & Cannon Co., Muskegon, Mich.

² American Cast Iron Pipe Co., Birmingham, Ala.

If we want to tap at 3000 degrees Fahr., we know that we must keep the current on a certain number of minutes after the smoke changes from brown to white. Our initial carbon and silicon must depend on how long we are going to heat the iron after that point.

WRITTEN REPLY BY THE AUTHOR

CLYDE L. FREAR: The author wishes to point out once more that this paper was written to describe the method of producing so-called soft iron in the electric furnace. Some of the procedure is described as a method of preventing difficulties which may occur at infrequent intervals due to some unconsidered variable entering into the melting operation. It must be considered that these difficulties are not common, but they must be kept in mind as it is much easier to prevent than to cure them after they have occurred.

The discussion by Mr. Austin and Mr. MacKenzie both bear out the author's statement that a well built bottom is necessary for close control of the carbon and silicon. As stated in the paper, there is always a tendency, with an acid lining, for the silicon to increase and the carbon to decrease, which tendency is increased by an increase in the temperature of the bath. It is the author's experience that the temperature of the bath should always be kept as low as possible, consistent with the properties required in the finished castings. The addition of stable carbide forming elements such as chromium will prevent to some extent this tendency to silica reduction as there will be less carbon available to cause this reduction.

A Study Of Six Bearing Bronzes

By OSCAR E. HARDER,* COLUMBUS, OHIO, AND
CARTER S. COLE,** NEW YORK.

Abstract

Six bearing bronzes of the compositions 80-10-10-0, 88-10-0-2, 83-7-7-3, 85-5-9-1, 70-10-20-0 and 70-5-25-0 per cent copper, tin, lead, and zinc, respectively, have been studied. The specimens were prepared according to current practices in the foundries of the four coöperating companies. Foundry practices are given. The data include tensile and yield strengths, elongation, reduction of area, modulus of elasticity, hardness at temperatures up to 450 degs Fahr. (232 degs. Cent.), resistance to compression, density as cast and after compressing at 100,000 lbs. per sq. in., and deformation by repeated pounding at room temperature and at 350 degs. Fahr. (177 degs. Cent.)

INTRODUCTION

1. To obtain the best possible bearing for a given use, one must fit to the particular service in hand the metallurgical properties of the shaft and bearing, the properties of the lubricant and the mechanical design and workmanship of the bearing and lubricating system. Mechanical design without doubt is the most important and the one on which least effort apparently is being put. Nevertheless, the properties of the bearing metals, as distinct from those of bearings, need to be thoroughly known in any comprehensive program looking toward improvement.

2. The information called for by specification and commonly determined by metallurgists on these properties of even ordinary bronze bearing alloys, has been incomplete and discordant in spite of many years of use. The engineer seeking information finds

* Asst. Director, Battelle Memorial Institute.

** Copper and Brass Research Association.

NOTE: This paper was presented before the Nonferrous Division session of the 1933 Convention of the American Foundrymen's Association.

Table 1
ALLOYS STUDIED AND MANUFACTURERS

Alloy No.	Nominal Composition				Furnished by Foundry			
	Cu	Sn	Pb	Zn	A	B	C	D
1	80	10	10	..	x	x	x	
2	88	10	..	2		x		x
3	83	7	7	3	x			x
4	85	5	9	1			x	x
5	70	10	20	..	x		x	
6	70	5	25	..		x		

Table 2
CHEMICAL COMPOSITION AND POURING TEMPERATURE
(MANUFACTURERS' DATA)

Alloy No.	Average Chemical Composition							Casting Temperature Degrees Fahr.			
	Cu	Sn	Pb	Zn	Fe	Ni	Sb	P	Max.	Min.	Av.
1	79.60	9.91	9.81	0.29	0.05	0.18	0.28	0.04	2175	1920	2020
2	88.00	9.91	0.25	1.72	2120	1900	1970
3	82.48	6.94	7.14	2.73	0.24	0.30	0.04	2160	1950	2080
4	84.60	5.36	8.61	0.82	0.04	0.14	0.34	0.04	2130	1890	2040
5	69.30	9.76	20.10	0.62	0.14	0.03	2090	1970	2020
6	70.67	4.78	24.37	Tr	0.07	2040	1860	1950

difficulty in getting reliable data on the alloys made by modern foundry practice.

3. In order to give him something more reliable, the Copper & Brass Research Association, with the cooperation of the Non-ferrous Bearing Metal Research Group,¹ has carried out at the Battelle Memorial Institute the work reported herein. We realize that these data constitute but one small step toward the development of improved bearings and believe that makers and users of bearings should jointly study the problems of engineering design, and of choice of materials plus design, in a more thorough fashion than has yet been done.

ALLOYS SELECTED

4. The alloys chosen for test were selected with two thoughts in mind, the importance of the alloy, and the need for the data. The tests made included tensile strength, compression, density,

¹Previous to the study at Battelle Memorial Institute, E. M. Staples had prepared two reports for the Nonferrous Bearing Research Group, in which he summarized the data available for various bearing bronzes and the specifications for such materials. These reports brought out the inconsistencies in the properties being reported and the lack of sufficient data on certain types of bearing bronzes. The Nonferrous Bearing Research Group consisted of the following members: Detroit Aluminum & Brass Corp., Bohn Aluminum & Brass Corp., Johnson Bronze Co., Bunting Brass & Bronze Co., G. H. Clamer, Fredericksen Co., Federal-Mogul Corp., and Lumen Bearing Co.

Table 3
FOUNDRY PRACTICE FROM MANUFACTURERS' DATA

Alloy No.	METAL CHARGE (percent)			Back Stock.	Type of Furnace.	Permeability.	MOLD DATA			FLUX OR COVERING			Shakeout Time, hrs.		
	Foundry.	Virgin Metal.	Secondary Metal.				Comp. Ingot.	Baled Scrap.	Borings.	Moisture, %.	Green Shear, (lbs. per sq. in.)	A.F.A. Bond, %.	Flux in Melting.	Flux in Ladle.	
1 A	50	..	20	30	Cupola	42	5.5	None	Borax	1/2
1 B	50	50	Electric	10	7.2	1.8	..	Borax	Charcoal	1/2
1 C	2	4	..	34	25	..	35	Crucible	14	7.0	2.0	7.6	Charcoal	..	7/8
2 D	40	60	Crucible	25	6.1	1.6	6.3	Charcoal	..	14
2 B	85	15	Electric	11	7.4	1.8	..	Borax	Charcoal	1/2
3 A	40	..	30	..	30	Cupola	42	5.5	None	Borax	1/2
3 D	4	..	80	16	Rotary	25	6.1	1.6	6.3	None	None	14
4 C	2	5	..	37	27	..	29	Crucible	14	7.0	2.0	7.6	Charcoal	..	14
4 D	18	82*	Rotary	25	6.1	1.6	6.3	None	..	14
5 C	12	21	..	67	Crucible	14	7.8	2.0	7.2	Charcoal	..	14
5 A	50	..	25	..	25	Cupola	40	5.0	None	Borax	1/2
6 B	50	50	Electric	10	7.9	2.2	..	Borax	Charcoal	1

NOTE: All except Foundry B deoxidized melts with phosphor-copper.

*In lot 4 D the Back Stock included over 20 per cent of bearing shells, from which the babbitt had been removed by "sweating out." The resulting antimony content was 0.45 per cent.

hardness and pounding. Test bars were furnished by four different foundries, each making up test specimens of three alloys.² Table 1 shows the nominal composition of the six alloys and the foundries furnishing each.

Chemical Composition

5. The average chemical composition of the various lots "as cast," together with casting temperature, are shown in Table 2.

Foundry Practice

6. Each company was left to choose its own practice, but was requested to report in detail the foundry practice used. This is summarized in Table 3.

Effect of Foundry Practice

7. The limited number of tests made did not show any one method of melting as superior to the others. Some small differences were noted, but there was no consistent relation regarding the effect of variations in foundry practice, except as noted under "Shake-out Time" below (see paragraph 21).

²The specimens were cast by Bohn Aluminum & Brass Corp., Bunting Brass & Bronze Co., Detroit Aluminum & Brass Corp., and Federal-Mogul Corp.

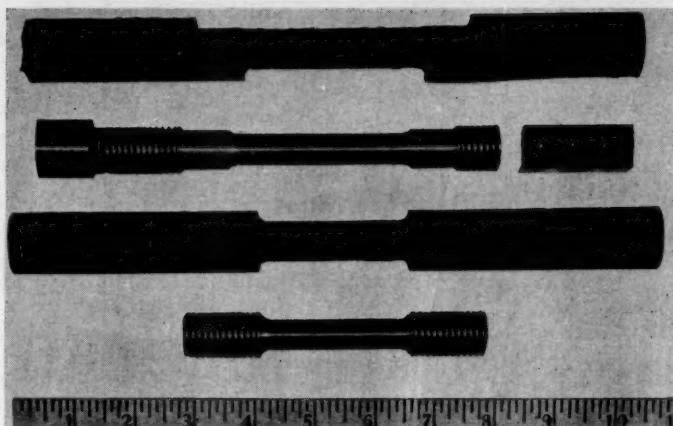


FIG. 1—CAST SPECIMENS AND THE TENSILE, HARDNESS AND COMPRESSION TEST SPECIMENS.

Table 4
TENSILE PROPERTIES

Alloy*	"As Cast"		MACHINED SPECIMENS—BATTELLE DATA**					
	Tensile Strength, lbs. per sq. in.	Elong., % in. 2 in.	Tensile Strength, lbs. per sq. in.	Yield Strength, lbs. per sq. in.	Elong., % in. 2 in.	Reduction of Area %	Modulus, $\times 10^{-6}$	
1	Av.	39,200	22.5	39,800	17,400	29.8	27.2	10.5
	Max.	43,400†	26.5†	44,000	19,300	41.0	34.4	12.2
	Min.	36,200	18.7	36,600	15,900	23.0	20.6	9.5
2	Av.	49,300	37.5	48,400	17,300	47.9	41.0	11.8
	Max.	50,400	40.6	50,600	18,000	58.0	42.5	12.2
	Min.	46,600	31.0	46,500	16,600	40.5	39.1	11.1
3	Av.	33,500	21.8	33,900	15,800	23.2	23.8	10.4
	Max.	35,600	25.0	35,800	16,500	27.5	24.8	11.4
	Min.	30,700	17.1	31,600	15,300	19.5	20.6	9.5
4	Av.	28,700	13.8	30,400	14,700	16.6	20.2	10.9
	Max.	32,700	20.3	34,000	15,300	22.0	25.1	12.8
	Min.	24,300†	7.5†	27,200	13,800	12.5	17.7	9.5
5	Av.	30,600	12.0	31,300	15,100	15.9	14.6	9.0
	Max.	32,200	12.5	33,700	16,200	17.0	16.3	9.4
	Min.	28,700†	11.5†	28,900	14,000	14.0	11.5	8.3
6	Av.	21,000	16.0	21,100	10,100	15.7	15.7	7.0
	Max.	21,800	18.0	21,800	10,200	17.5	17.0	7.4
	Min.	20,100	14.0	20,000	10,000	13.5	14.2	6.7

†Tensile test on 0.505-in. diameter machined specimen. All other "as cast" data on rough specimens of about $\frac{1}{2}$ -in. diameter in the reduced section. (Manufacturers' Data)

**Battelle Data on machined specimens of 2-in. gage length and 0.505-in. diameter.

*The types of alloys examined, corresponding with the alloy numbers shown in the table, were as follows: Alloy No. 1, 80-10-10; No. 2, 88-10-0-2; No. 3, 88-7-7-3; No. 4, 85-5-9-1; No. 5, 70-10-20; No. 6, 70-5-25.

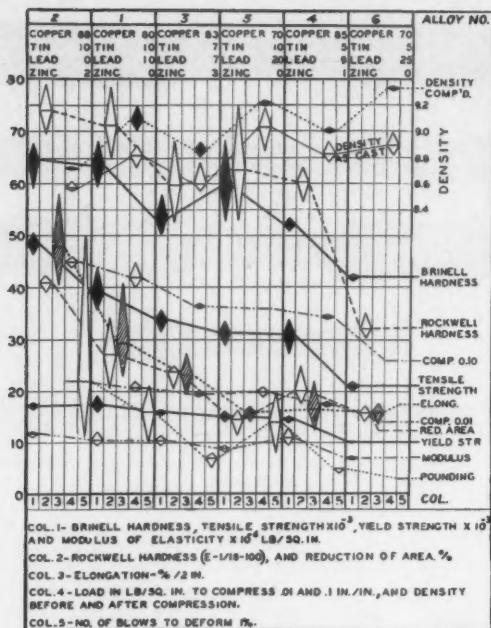


FIG. 2—PROPERTIES OF SIX BEARING BRONZES. ALLOYS ARRANGED WITH DECREASING STRENGTH FROM LEFT TO RIGHT.

Test Specimens

8. All castings were fin-gate type,³ although there was a slight modification in those furnished by one of the companies. Specimens are illustrated in Fig. 1, which also shows the form and position in the ingot of the compression test specimens (on the left) and the hardness and pounding test specimens (on the right). All specimens were gated for a length of 3½ inches on each end.

Tensile Tests

9. The tensile specimens as shown in Fig. 1 were 0.505 in. in diameter with a 2-in. gage length and threaded ends. They were selected to represent one of the first molds poured, one near the middle and one near the last of each melt. The testing was done with a 72,000-lb. Amsler machine. Elongation was read with

³A.S.T.M. Standards, 1930, pp. 678-681.

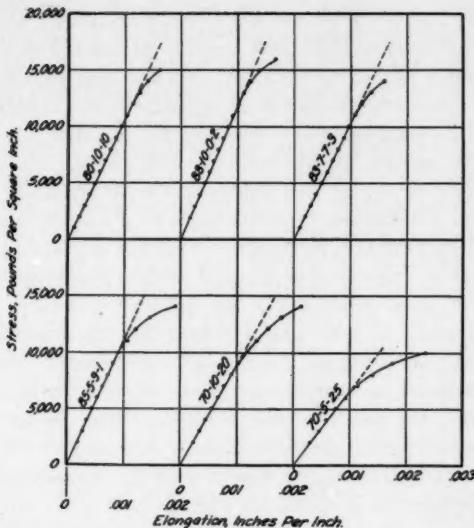


FIG. 3—REPRESENTATIVE STRESS-STRAIN CURVES FOR SIX BEARING BRONZES.
(SEE TABLE 4 AND FIG. 2 FOR FURTHER DATA.)

a Goodyear-Zeppelin Extensometer—readings to 0.00001 in. For the specimens of the 80-10-10⁴ alloy, elongation was measured on a 1-in. gage length with a Huggenberger Tensometer.

10. The summary of the tensile tests is given in Table 4 and is graphically shown in Fig. 2. Each manufacturer made tensile tests on "as cast" specimens, and there was very good agreement between these values as reported and those obtained by the test of the machined specimens at Battelle. The values of the percentage of elongation "as cast" were lower, however, averaging about 80 per cent of those values obtained for the machined specimens.

Yield Strength

11. The yield strengths were determined from the stress-strain curves, using A.S.T.M. tentative specification E 6-32 T⁶ and taking as the yield strength a set of 0.001 in. per inch or 0.1 per cent. The values are given in Table 4 and shown graphically in Fig. 2. The stress-strain curves are shown in Fig. 3.

⁴In indicating the composition of the alloys, the amounts of the metals are given in this order: copper, tin, lead, zinc.

⁵Proceedings A.S.T.M., v. 32, pt. 1, 1932, pp. 967-969.

Compressive Properties

12. For compression tests, specimens having a diameter of 0.944 in. and a height of 0.850 in. (90 per cent of the diameter) were used. Loads were increased by increments of 2,000 lbs. per sq. in. up to a load of 20,000 lbs. per sq. in., then by increments of 10,000 lbs. per sq. in. up to 50,000 lbs. per sq. in., then loaded to 75,000 lbs. per sq. in. and finally to 100,000 lbs. per sq. in. All stresses were calculated on the original diameter of the specimens. The specimens were measured with a surface gage reading to 1/100,000 in., which was checked with Johannsen gages.

13. For each reading the specimen was loaded to the desired point and the load held constant for 30 seconds, and then released and the height of the specimen measured. The load permanent-deformation curves were then plotted, using a semi-logarithmic scale for the decrease in height. The results are shown graphically in Figs. 4 and 5 and also given in Table 5. The correlation of the compressive properties of the bronzes studied with the other properties is shown in Fig. 2.

Density

14. Density determinations were made of all the alloys in-

Table 5
DENSITY AND COMPRESSION TESTS

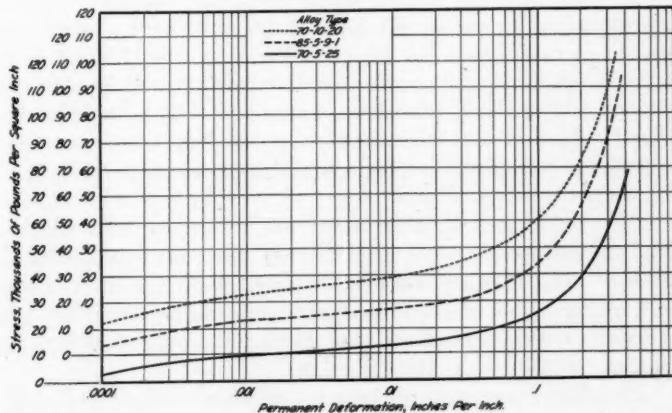
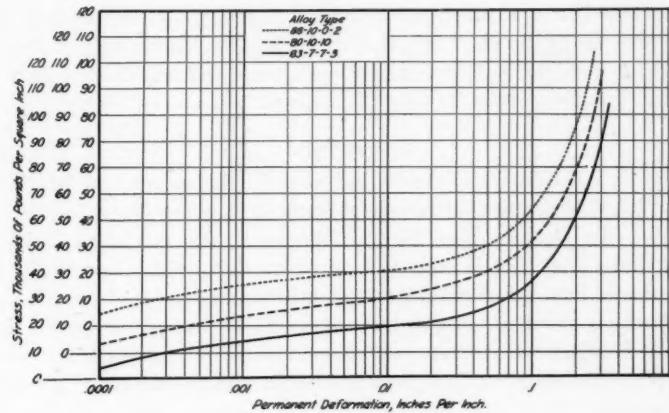
Alloy No.	Nominal Composition	Density at 24 degs. Cent.			Compressive Properties		
		As Cast.	Compressed.	Change in Volume (2)	0.001 in.	0.01 in.	0.1 in.
1 80-10-10	Av.	8.820	9.080	-2.98	14.0	20.7	42.0
	Max.	8.900	9.196	-3.60	15.0	22.0	45.0
	Min.	8.736	9.013	-2.32	13.0	20.0	40.0
2 88-10-0-2	Av.	8.573	8.723	-1.72	16.0	22.0	45.0
	Max.	8.626	8.737	-1.98	16.0	22.0	46.0
	Min.	8.551	8.712	-1.27	16.0	22.0	44.0
3 83-7-7-3	Av.	8.604	8.864	-2.93	14.0	19.5	36.5
	Max.	8.771	8.926	-3.58	14.0	20.0	37.0
	Min.	8.543	8.813	-1.75	14.0	19.0	36.0
4 85-5-0-1	Av.	8.831	9.006	-1.95	14.5	17.5	34.5
	Max.	8.930	9.037	-2.56	15.0	18.0	35.0
	Min.	8.772	8.993	-1.19	14.0	17.0	34.0
5 70-10-20	Av.	9.037	9.217	-1.96	14.5	20.0	41.0
	Max.	9.184	9.242	-4.07	15.0	21.0	46.0
	Min.	8.852	9.192	-0.38	14.0	19.0	36.0
6 70-5-25	Av.	8.898	9.334	-4.67	10.0	14.0	26.0
	Max.	8.995	9.346	-5.50	10.0	14.0	26.0
	Min.	8.832	9.316	-3.68	10.0	14.0	26.0

¹After load of 100,000 lbs. per sq. in.

²In per cent of original. All show decrease in volume, or increase in density.

cluded in this study, using the compression test specimens. Density before compression was determined from the calculated volume of the specimen and its weight; after compressing, by the Archimedes principle. The results are given in Table 5; see also Fig. 2.

15. The significant observation with reference to the densities is that in all cases the specimens were more dense after being subjected to a compression of 100,000 lbs. per sq. in. Cold working, as by compression, decreases the density of annealed sound



FIGS. 4 (ABOVE) AND 5 (BELOW)—MEAN STRESS-PERMANENT DEFORMATION CURVES FOR SIX BEARING BRONZES.

specimens of pure metal. That this does not hold for these alloys seems to show that the specimens "as cast" are somewhat porous, and that compression closes up the pores and makes them more dense.

16. There was good agreement for the values of the specimens of the same type of composition except in the 70-10-20 type, where the Lot C specimens were sufficiently porous to affect noticeably the tensile strength and hardness.

17. These results would suggest the possibility of using the change in density, or the decrease in volume, of specimens during compression as a means of detecting porosity. It seems that excessive porosity could be detected in this way, and such a determination might be useful in research work and might even have a place in the routine testing of castings.

Hardness

18. Hardness is rarely included in American specifications for bearing bronzes, although in some it is given for information only. In German specifications, however, the minimum Brinell hardness generally is included. As regards specifications, while it seems to have been considered of minor importance from an engineering viewpoint, hardness is one of the important properties of a bearing. Hardness, not only at room temperature but also at elevated temperatures encountered in operation, should be considered.

19. Hardness of the six alloys was tested at intervals from room temperature up to 450 degs. Fahr. (232 degs. Cent.). At room temperature both the Rockwell hardness ("E" scale) $\frac{1}{8}$ -in. ball and 100-kg. load, and the Brinell hardness using a 10-mm. ball and 500-kg. load, have been determined. The Rockwell and Brinell values of the hardness of the different bronzes at room temperature are given in Table 6 and are shown graphically in Fig. 2.

20. Only the Brinell hardness values were determined at elevated temperatures. The summarized data are also given in Table 6. The hardness decreased slightly with increase in temperature up to 350 degs. Fahr., but in the range from that temperature to 450 degs. Fahr. the hardness remained practically constant, or showed a slight increase.

21. There appeared to be some relation between the shake-out time and the hardness; this relation was not always consistent, however, and definite conclusions cannot be drawn without

Table 6
HARDNESS TEST DATA

Alloy No.	Nominal Compo- sition	Rockwell*			Brinell†		
		75°F.	75°F.	150°F.	250°F.	350°F.	450°F.
1	80-10-10	Av.	71	64	61	59	56
		Max.	79	68	66	63	60
		Min.	65	59	58	56	53
2	88-10-0-2	Av.	74	66	64	60	58
		Max.	79	70	68	62	60
		Min.	69	59	61	55	54
3	83-7-7-3	Av.	62	53	52	52	50
		Max.	68	58	60	56	53
		Min.	52	49	44	49	46
4	85-5-9-1	Av.	60	52	52	50	49
		Max.	63	53	57	52	50
		Min.	56	51	46	44	45
5	70-10-20	Av.	63	60	58	57	53
		Max.	75	69	66	63	61
		Min.	53	53	51	52	48
6	70-5-25	Av.	32	42	41	40	38
		Max.	35	43	42	41	38
		Min.	30	41	41	38	38

*Rockwell Hardness ("E" Scale), $\frac{1}{16}$ -in. ball and 100-kg. load.

†Brinell Hardness—10-mm. ball, 500-kg. load, 1 minute.

making a series of tests in which the shake-out time is varied for each type of alloy. In the 70-10-20 alloys the specimens which were shaken out in a half-hour were definitely harder than those which remained in the mold for 14 hours. The same tendency was shown for the 88-10-0-2 and 83-7-7-3 alloys, but to a lesser degree. In the case of the 80-10-10 alloys, there was practically no difference in the hardness for variations of from one-half to $7\frac{1}{2}$ hours in the time in the mold.

22. Hardness test specimens were $1\frac{1}{8}$ in. long and $11/16$ in. square, prepared from the long end of the test bars (Fig. 1). Tests at elevated temperatures were made with a 10-mm. tungsten carbide ball. At room temperature (75 degs. Fahr.) a water bath was used; at 150 and at 250 degs. Fahr., a glycerin bath; at 350 and at 450 degs. Fahr., a bath of "Fisher Wax." The load was applied for one minute. Four diameters at 45 degrees apart were read on each impression, two tests made on each block, and three blocks from each manufacturer were tested.

Pounding

23. The importance of resistance of bearing bronzes to deformation by repeated pounding is obvious, and the results of pounding tests may be used to correlate other data in evaluating

the properties of these materials. Pounding tests were made both at room temperature and at 350 degs. Fahr., the latter approximating the maximum operating temperature under normal conditions.

24. Pounding tests were made using a machine similar to that used on bearing metal research at the Bureau of Standards.⁶ Test conditions are similar and results comparable. The machine is constructed to give repeated blows in compression from a pile-driver-like hammer weighing 7.15 lbs. and falling from a height of 2 inches. An auxiliary device catches the hammer on the rebound to prevent uncontrolled strokes.

25. Small cylinders of bronze 10 mm. in diameter and 20 mm. long were used as test specimens. A small hole was drilled into the base to locate the specimen on a pin on the anvil of the pounding machine. Readings were made to the nearest 0.005 mm. and the number of blows between readings regulated to provide sufficient points to draw a smooth curve of the test results thus obtained.

26. The testing at 350 degs. Fahr. (177 degs. Cent.) was carried on in a similar manner, care being taken to maintain a uniform temperature in the specimen and the anvil. Deformation was read by means of a telescope and comparator, the top and bottom of the specimen being read each time. Twenty minutes was allowed after the specimen had been placed in the furnace before pounding was started.

27. The alloys showed resistance to pounding at room temperature and at the elevated temperature, for both 1 per cent and 5 per cent deformations, in the following order:

Order of Resistance to pounding.	Nominal Composition.	Tin/Copper Ratio.
1	88-10-0-2	0.114
2	80-10-10	0.125
3	70-10-20	0.143
4	83-7-7-3	0.084
5	85-5-9-1	0.059
6	70-5-25	0.067

28. These showed quite definitely that with an allowance for lead content, alloys show resistance to pounding as a function of the tin/copper ratio. It was noted that those specimens which had been allowed to remain in the mold for a long time before shaking out, showed definitely a lower resistance to pounding in some alloys. The general results of the pounding tests are sum-

⁶H. J. French and others, *Bureau of Standards Jl. of Research*, v. 1, 1928, pp. 343-421, and *Research Paper No. 13*.

Table 7

RESULTS OF POUNDING TESTS AT ROOM AND ELECTRIFIED TEMPERATURES

Alloy No.	Nominal Compo- sition.	No. of Blows to Produce Deformation of:				Percent Deformation at End of Test. ¹	
		1% 75°F.	5% 75°F.	1% 350°F.	5% 350°F.	75°F.	350°F.
1	80-10-10	Av. 15	960+	6	850+	4.50	4.30
		Max. 21	1000+	8	1000+	5.20	7.20
		Min. 10	800	4	250	3.60	2.50
2	88-10-0-2	Av. 22	900+	7	900+	3.95	4.05
		Max. 50	1000+	9	1000+	5.55	6.85
		Min. 11	630	5	320	2.45	2.15
3	83-7-7-3	Av. 6	220	3	38	8.35	8.35 ²
		Max. 8	600	4	48	10.05	9.50 ³
		Min. 5	100	3	29	5.35	7.15
4	85-5-0-1	Av. 5	54	2	23	9.65 ⁴	9.60 ⁵
		Max. 5	70	3	26	10.35 ⁶	10.35 ⁷
		Min. 4	34	2	19	7.65	8.60
5	70-10-20	Av. 13	750+	4	560+	5.10	5.20
		Max. 20	1000+	7	1000+	6.75	5.95
		Min. 8	280	3	175	3.25	4.10
6	70-5-25	Av. 3	23	2-	15	10.55 ⁸	10.20 ⁹
		Max. 3	23	2-	15	10.75 ⁸	10.45 ⁹
		Min. 3	22	2-	15	10.40 ⁸	10.05 ⁹

¹This deformation, except as noted, is for 1000 blows, the maximum given any one specimen.

²For average of 930 blows per specimen.

³For 1000 blows; another specimen had 8.7 per cent for 800 blows.

⁴For average of 800 blows.

⁵For average of 575 blows.

⁶For 400 blows; also another with 10.65 per cent for 800 blows.

⁷For 260 blows; also another with 10.45 per cent for 400 blows.

⁸For 200 blows.

⁹For 60 blows.

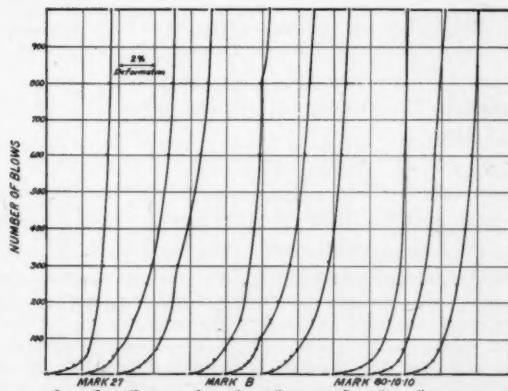


FIG. 6—ROOM TEMPERATURE POUNDING TEST CURVES FOR 80-10-10 ALLOYS, SHOWING RELATION BETWEEN NO. OF BLOWS AND PER CENT DEFORMATION OF SPECIMEN. WEIGHT OF HAMMER, 7.15 LBS.; FALL, 2 INCHES.

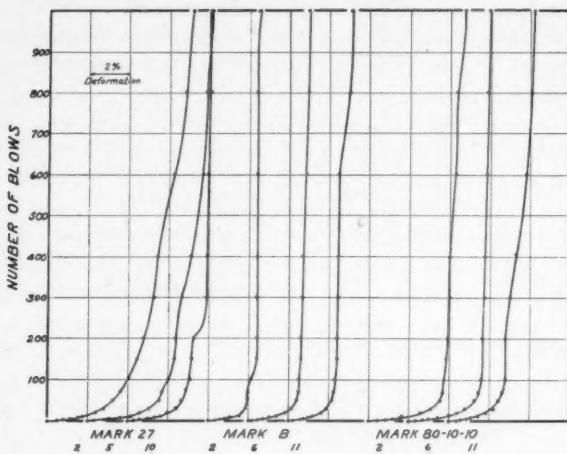


FIG. 7—ELEVATED TEMPERATURE (350 DEGS. FAHR., 177 DEGS. CENT.) POUNDING TEST CURVES FOR 80-10-10 ALLOYS, SHOWING RELATION BETWEEN NO. OF BLOWS AND PER CENT DEFORMATION OF SPECIMEN. WEIGHT OF HAMMER, 7.15 LBS.; FALL, 2 INCHES.

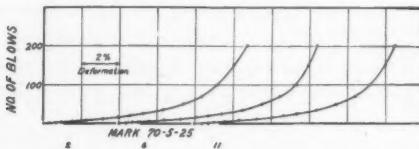


FIG. 8—ROOM TEMPERATURE POUNDING TEST CURVES FOR 70-5-25 ALLOY, SHOWING RELATION BETWEEN NO. OF BLOWS AND PER CENT DEFORMATION OF SPECIMEN. WEIGHT OF HAMMER, 7.15 LBS.; FALL, 2 INCHES.

marized in Table 7, while typical pounding curves are illustrated in Figs. 6, 7 and 8.

29. Fig. 6 shows the pounding test curves for the 80-10-10 alloy at room temperature. These curves represent tests on three specimens from each of three manufacturers. Similar tests at 350 degs. Fahr. (177 degs. Cent.) are shown in Fig. 7. The effect of a high-lead content on the pounding test curves is shown in Fig. 8, which presents curves for the 70-5-25 alloy tested at room temperature.

30. It will be noted that with the exception of the alloy 70-5-25, all of the curves show that the deformation proceeds in an irregular manner. The specimens may deform only one-tenth or two-tenths per cent in 300 blows and then deform two-tenths in

the next 10 blows. This probably is due to a certain amount of work hardening, followed by a new range of deformation.

31. Resistance to deformation at elevated temperatures was lower for small amounts of deformation, but roughly the same for the deformations of around 5 per cent and those produced by 1,000 blows. Thus, at elevated temperatures the alloys show somewhat less resistance to small deformations but work-harden to the same degree as at room temperature. It also shows that the temperature used (350 degs. Fahr.) is below the rapid annealing temperature of the alloys studied.

SUMMARY AND CONCLUSIONS

32. Six bearing bronzes representative of modern practice in four foundries have been tested for tensile and yield strength, elongation, reduction of area, hardness at room temperature and at elevated temperatures, resistance to pounding at room temperature and at 350 degs. Fahr. (177 degs. Cent.)

33. Good agreement has been found between the properties of specimens of the same alloys cast in the different participating foundries.

34. Such properties as tensile strength, elongation, and reduction of area were found to be well above present specification requirements for the alloys tested.

35. It is possible to prepare data sheets from the results obtained for these six bearing bronzes which will give helpful engineering information.

36. It appears that similar data are needed for other extensively-used bearing bronzes.

37. There is need for correlation between the properties reported in this investigation and actual bearing performance in different types of service.

DISCUSSION

G. H. CLAMER:¹ I am told that steel-backed bearings are becoming more popular than bronze-backed bearings, particularly in the automotive field. I was talking with a friend about the design of the present engines used in the automotive industry, and particularly about the changes that have been made in the design of bearings. The tendency is to use smaller bearings. That is possible as a result of forced lubrication. Can anybody give authentic information as to whether the trend is continuing in the direction of steel-backed bearings?

¹ Ajax Metal Corp., Philadelphia.

C. S. COLE: During the past year there has been little change from bronze to steel-backed bearings. Certain groups of manufacturers still tend toward the steel-backed bearing; some because they think bronze-backed bearings are not strong enough and others because they may get a lower price on the steel-backed variety. The alloys that we tested, especially the 80-10-10, show that some bronze bearing alloys have much higher properties than are called for in present specifications. A careful inspection of the data presented in our paper should make the users of steel-backed bearings again think of bronze and make those contemplating a change from bronze to steel reconsider.

CHAIRMAN SAM TOUR:² Mr. Cole's statement that some of these alloys show much higher properties than specifications require, is interesting. I think a number of nonferrous specifications indicate that one alloy is better than another, whereas in actual practice, the reverse is true. To that extent, I believe some nonferrous producers have been at fault in trying to keep specifications too low as far as physical properties are concerned. That allows makers of competing metals to point out to users how poor some copper-base alloys are, based on specifications, when the specifications do not indicate really what the alloy will do.

MEMBER: At our factory, they have experimented with steel-backed bearings and found many objections. The idea of studying that type bearing was the question of price. Many users of bronze have desired to get a quality article at a little bit less than they should be asked to pay for it. The idea was developed by two or three other concerns and was introduced as a cure-all proposition. It was possible for users to buy from a jobbing house without any definite knowledge of application, practicability, correct design or proper lubrication. There has been a decided set-back to the use of steel-backed bronze which is not due entirely to the material itself or design, but rather to the fact that it has been applied or used improperly. However, if prices keep increasing, I would not be surprised to see a new demand created for the steel-backed bronze bearing.

G. H. CLAMER: In referring to a bronze steel-backed bearing, do you refer to the lead-copper alloy?

MEMBER: I was referring to a copper-tin-lead bronze rather than babbitted steel.

CHAIRMAN TOUR: Mr. Clamer's original question was on the substitution of steel for bronze in the backing itself, in a babbitt-lined bearing.

G. H. CLAMER: Yes, but the tendency now is to use a steel back with a copper-lead alloy lining, which ordinarily is referred to as bronze. It is a 75 per cent copper and 25 per cent lead or 60 per cent copper and 40 per cent lead alloy, eliminating the babbitt. Those bearings are used extensively.

J. H. ROAST:³ I was in Glasgow, Scotland, at the Beardmore company a year or two ago and that company was recommending strongly a combination of all three, a steel back, a tin-lead bronze and a babbitt top.

² Lucius Pitkin, Inc., New York.

³ McGill University, Montreal, Canada.

C. S. COLE: I think that it is important that we note how much above the specifications good foundry practice can make these standard bearing metals. The specifications may be slightly low but probably good foundry practice can make bearing bronzes that will come within the limits of those discussed in the paper.

E. F. HESS:⁴ If the time comes when there is a definite test bar for brass alloys, it will be possible to talk the same language. A lot of test bars are made under varying conditions, shape, etc., and they give quite a variation in tensile strength.

G. H. CLAMER: Mr. Hess' statement is to the point. The tests, I gather, were made with the web or fin-gate type bars that were not machined except in a few cases.

C. S. COLE: The preliminary tests made by the foundries were made on rough castings, except as noted in Table 4. All tests at Battelle Memorial institute were made on machined specimens.

G. H. CLAMER: These tests show, for the 80-10-10 alloy, as high as 44,000 lbs. per sq. in. As a matter of fact, if you cut a test piece from a bearing, such as those used on railroad cars or locomotives, it is hard to get anything above 20,000 to 22,000 lbs. per sq. in. for that alloy. Section exerts an influence on physical properties. In the specification we are indicating what might be expected in a commercial casting instead of the results obtained from bar tested under ideal conditions.

J. H. ROAST: I think that the idea of increasing the specifications is not good, I have found a growing tendency among Canadian engineers to say, "Yes, this is your specification on your test bar, but what is it in the design that we are discussing?" In my experience that attitude is growing and if it keeps increasing, we are better off if we have specifications more in line with results that may be obtained under a reasonably practical application.

C. S. COLE: This discussion leads back to the old controversial subject of the value of test bars. Our data presents the values obtained from test bars prepared in accordance with the A.S.T.M. specifications.

MEMBER: With reference to the test results, there are so many conditions that can affect the finished casting that great care should be exercised before deciding that the results of these particular tests are complete. My company has gone through an experience lately that brought forth that question. I was noticing density in the development of a bronze for seal rings. By pouring at a slightly higher temperature it was possible to fume-off a little of the tin and by that process develop a much denser bronze than the same analysis under ordinary foundry practice. Density, hardness, etc., can be affected by foundry practice so radically that the foundrymen must be careful before determining that these results cover the definite points that have been brought out.

⁴ Ohio Injector Co., Wadsworth, O.

Beryllium-Copper Castings—Foundry Practice, Heat Treatment, Properties

BY EDWIN F. CONE, NEW YORK

Abstract

This is a progress report on developments in the production of beryllium-copper alloy castings—a new field. Source of the metal beryllium, some of its properties, and its recovery from the ore as a "master alloy" of 12.5 per cent Be and the balance pure copper, are outlined. Data thus far indicate that the foundry practice closely resembles that for phosphor bronze. The master alloy is used as a raw material for introducing beryllium into the alloy, the aim being to secure a range of 1.50 to 2.50 per cent Be. The 2.50 beryllium-copper alloy is regarded as standard for castings. Heat treatment is interesting, the alloys belonging to the precipitation-hardening class. After a solid solution treatment, precipitation hardening can be attained by heating 2 to 3 hours at 525 to 575 degs. Fahr. and air cooling. Very high tensile strengths result, with hardness values up to 400 Brinell. Properties of the rolled or wrought alloys of 2.25 per cent Be have even more excellent properties than the cast products; some are given in the paper. Electrical and thermal conductivity are discussed. Data included show the hardness range for certain compositions heated at 525 to 575 degs. Fahr. for varying periods. That the alloys of beryllium and copper have high wear-resisting properties is demonstrated by results of some authoritative tests. Anti-corrosive properties have been demonstrated by certain applications in the oil industry where sulphur and abrasive conditions prevail. Applications of the 2.50 per cent beryllium-copper alloys are dealt with in concluding paragraphs. A bearing sleeve for a new adjustable-pitch propeller has been made in fairly large numbers, precipitation hardened before machining. Racing-boat propellers of beryllium bronze have

NOTE: This paper was presented and discussed before one of the nonferrous sessions of the 1933 Convention of the American Foundrymen's Association.

also been made. Other products include cams, gears and pinions, etc. A nickel-beryllium master alloy of about 12.5 per cent Be also is available, with some interesting products believed possible. The whole field of beryllium and its non-ferrous alloys has only been scratched

1. A comparatively new alloying element has been recently introduced to metallurgists—it is beryllium (Be), originally named glucinum. Some of the properties which it bestows as an alloy are exceedingly interesting and are attracting wide interest in several industrial fields. The development is in its early stages, but enough investigation and progress have been made in this country and in Europe to warrant the prediction that the metal will play an important role in industry.

2. Beryllium resembles the aluminum-magnesium family in some of its properties. It is one-third lighter than aluminum and one-sixteenth heavier than magnesium. Put in another way, it is about 35 per cent lighter than aluminum, having a specific gravity of 1.85 and a melting point of about 2335 degs. Fahr.

3. Within the memory of many now living, aluminum was as unfamiliar some fifty years ago as beryllium is now. About 25 to 30 years ago the same was true of magnesium. Today these two metals are essential factors in the engineering and industrial life of the world, either as metals or as alloys in the wrought or cast condition. It is not improbable that, in the not-distant future, beryllium may assume a place similar to aluminum and magnesium in industry.

4. Like aluminum and magnesium, beryllium occurs in commercial quantities in nearly every country. It exists more commonly as the ore, beryl, a beryllium-aluminum-silicate which contains on the average about 5 per cent beryllium.

5. Recovery from the ore is difficult and expensive on a small scale, which explains the present high cost of beryllium. It is extracted by the electrolytic treatment of a fused chloride or fluoride bath of the salts of the metal—similar to the aluminum or magnesium reduction processes. Unquestionably, as progress is recorded and demand for the metal increases, its cost will be lowered. This is the history of aluminum, magnesium, vanadium and molybdenum, all highly important in their various fields.

A Beryllium Master Alloy

6. Up to the present time the principal application of beryllium as an alloy has been in the nonferrous industry as both

wrought or fabricated and cast products. The commercial development has been confined to alloys of copper, but other series of alloys are under investigation. For this purpose a so-called "master alloy," containing about 12.5 per cent beryllium with the balance copper, has been prepared commercially and is being marketed by an American company. With this master alloy as a raw material base, varying percentages of beryllium are being introduced into copper with about 1.50 to 2.50 per cent beryllium in the prevailing alloyed products.

7. Thus far, most of the research which has been accomplished has been in the fabricated or wrought alloys of beryllium and copper. In the castings field, research has not advanced as far. Considerable progress, however, has been made, and it is the purpose of this paper to present a progress report on some of the salient features.

8. Actual foundry production in this new field, on a somewhat limited scale, has been conducted by the American Manganese Bronze Co., Holmesburg, Philadelphia, where commercial orders for certain types of products of beryllium bronzes have been filled. Some experimental investigative work there has also been conducted, and the writer, who has visited this foundry, is indebted to this organization for some reliable data which are included in this paper.

How to Add the Beryllium

9. To introduce definite percentages of beryllium into a copper product, the 12.5 per cent master alloy is obtained from the producer, who furnishes the exact beryllium content with each shipment. This alloy is very brittle in its high state of concentration and can easily be broken up into suitable pieces for addition purposes. The calculation of the charge is made according to the following formula:

$$\frac{T \text{ Be}_2}{\text{Be}_1} = \text{pounds of master alloy.}$$

where T = total pounds of metal desired,

Be_2 = per cent beryllium desired in final product, and

Be_1 = beryllium in master alloy.

10. In producing a beryllium-copper alloy casting, the copper is dead-melted in a clay graphite crucible under a heavy covering of charcoal. An electric furnace can be used. The melt is deoxidized thoroughly with boronic copper, calcium, lithium,

silicon or magnesium. Practice of the American Manganese Bronze Co. is to use boronic copper. Beryllium has a great affinity for oxygen and of itself would completely deoxidize the copper; thus, in order to prevent consumption of beryllium, the above-mentioned deoxidants are preferably and necessarily used. Free boric acid should not be used, although glass borax in the charcoal covering may be employed as additional protection.

11. To offset the cooling effect produced by the addition of the master alloy, the bath of pure copper is heated to just above 2200 degs. Fahr. before additions. After the additions of small pieces of the 12.5 per cent alloy, the bath is stirred with a graphite, plumbago or nichrome rod, using special care so that the alloy does not float on top. Too active stirring is to be avoided because of the unfavorable effect due to absorption of gases. Iron stirring rods are prohibited.

12. The next point is to bring the bath by cooling or heating to the pouring temperature, which should be as low as possible—from 1900 to 2000 degs. Fahr. is recommended by the alloy producing company. The practice at the Holmesburg foundry is to keep the temperature for pouring as near 2000 to 2100 degs. Fahr. as possible, avoiding the necessity of reheating to achieve the desired result. Careful pyrometric control and good foundry practice have made this possible.

Pouring Temperatures and Foundry Practice

13. Casting of the alloy is then conducted according to usual methods. Pouring should take place as soon as practicable at the proper temperature after additions of the master alloy. Pouring too hot with relation to the job in hand should be avoided, and at too low a temperature unfavorable results follow. All pouring operations are conducted with a charcoal covering on the metal.

14. Sand casting practice is that usually followed for castings. For metal to be subsequently fabricated, chill molds are preferable. If cast iron molds are used they should be thoroughly cleaned, warmed to about 250 to 300 degs. Fahr. and dressed by a coating of soot obtained by customary methods.

15. As to sand casting in general, operations should closely approximate the practice ruling in the casting of phosphor bronze. The use of heads and gates is necessary to insure the usual even flow of metal in the mold.

16. Practice at Holmesburg, where heats of 200 lbs. are made, is to use green-sand molds. Heading and gating are simple, the

practice usually following that for aluminum or phosphor bronze. Size of heads and gates approximate those used for phosphor bronze and are only about one-fourth the size necessary for aluminum bronze; that is, the shrinkage is much less than for aluminum bronze, approximating that of phosphor bronze. Linear shrinkage has been found to be about 3/16 in. per foot for the 2.5 per cent beryllium alloy.

17. The melting loss on 200-lb. lots has been found to be about 9 per cent for both the beryllium and the copper. Rejections have not exceeded 2 per cent. Castings come out clean, smooth and sound in most cases and have a golden copper color. A feature is the high value of the scrap, from which the possible beryllium recovery is high.

Beryllium as a Deoxidizer

18. Because of its inherent deoxidizing power, beryllium is strongly recommended for this purpose in making copper castings, in addition to its use as an alloying agent. For this purpose, standard practice is to add 0.01 to 0.02 per cent beryllium, either alone or, preferably, following standard deoxidizers such as boron copper, to copper and copper-base alloys. Experience reveals the fact that the soundness of such castings, as well as their machinability and workability, will be measurably improved and clean, sound castings free from imperfections secured. If percentages of beryllium up to 0.05 to 0.10 per cent be employed, further improvement of machinability is claimed to result.

Interesting Heat Treating Practice

19. Heat treatment of the beryllium-copper alloys is very interesting and important. Only by carefully regulated practice can the full properties be brought out.

20. In general, the beryllium-copper after casting should be held at a maximum of 1475 to 1500 degs. Fahr. for at least two hours. This insures the complete solution of any precipitated beryllides. The mass and shape of the casting will determine the length of time required, which is also increased as the percentage of beryllium increases. Because the disintegrating temperature of beryllium-copper seems to lie somewhere around 1550 degs. Fahr. or slightly above, the temperatures of 1475 to 1500 degs. Fahr. should be carefully regulated by pyrometric control.

21. After soaking at this temperature, the product is quenched in cold water which should be accomplished as soon as possible,

after which it will be found to have its maximum softness, maximum elongation and minimum tensile strength. This treatment, readily recognized as the solid solution treatment, is essential and preliminary to the precipitation hardening treatment described later. It should in all cases be applied before machining. For a 2.5 per cent beryllium-copper alloy, the Brinell hardness of the quenched product is about 100. Such castings can be readily machined.

Hardening by Precipitation Treatment

22. A feature of the beryllium-copper alloys, or the beryllium bronzes, is their susceptibility to a precipitation hardening treatment, which greatly enhances their properties and value as engineering materials. This treatment is applied to the annealed or solid solution beryllium-copper castings. Recommended practice is heating at 525 to 575 degs. Fahr. and no higher for varying periods of time so as to insure complete precipitation. If a higher temperature is employed the precipitated beryllide tends to be re-absorbed, with a consequent decrease in hardness. A peculiarity of these alloys is that a slight decrease in volume results from the precipitation hardening treatment, which may be set down as approximately 0.001 in. per inch of diameter.

23. To obtain the expected strengths, hardnesses and wear-resisting qualities from this special treatment, for castings containing 2 per cent or over of beryllium the heating period is recommended as 3 hours. There is a progressive increase in electric conductivity between a 3-hour and a 24-hour treatment, with a maximum obtained from the 24-hour treatment. For castings containing 1.25 to 2 per cent beryllium a 5-hour treatment is recommended to insure desired characteristics. Further investigation may somewhat alter these recommendations.

Special Treatment to Avoid Distortion

24. A special procedure is suggested, by the corporation producing the alloy, for castings of thin section or complicated shape, because there is the possibility of some distortion taking place due to the weight of the casting pulling itself out of shape. To overcome this possible distortion, the following heat-treating sequence is suggested:

- (1) Heat in as non-oxidizing an atmosphere as possible at 1475 to 1500 degs. Fahr. (maximum) for 2 hours. Cool in air (normalizing).

- (2) Reheat, properly supported, and quench in water.
- (3) Rough machine, *i.e.*, leave enough material for a safe finishing grind.
- (4) Reheat, properly supported in a form, and requench.
- (5) Precipitation harden on the form.
- (6) Wet grind to a finish.

25. This practice may be varied, two steps (Nos. 2 and 5) that are essential being always included. Step No. 3 may be omitted, leaving all machining until after precipitation hardening, employing Nos. 1, 2, 4 and 5, or at least steps Nos. 2 and 5. Any grinding in the hardened state should be done wet. Machining can be accomplished with any good grade of high-speed steel.

When Pickling Is Necessary

26. If pickling is found necessary, a 5 to 7 per cent solution of nitric acid, a 10 per cent solution of sulphuric acid, or a 10 per cent solution of nitre cake may be used. This should be carried out one step prior to the last heat treatment, which will, of course, eliminate any occluded hydrogen caused by pickling.

27. The alloy which has received the most attention thus far and which has been found to be best suited for certain castings is the one containing about 2.50 per cent beryllium. In the precipitation hardened condition (after 3 hours heating at 575 degs. Fahr.) the properties as given by the alloy producer are as follows:

Tensile strength, lbs. per sq. in.	110,000 to 120,000
Yield point, lbs. per sq. in.	85,000 to 90,000
Elongation in 2 in., per cent	0.0
Reduction of area, per cent	1.0
Brinell hardness	375 to 400
Electrical conductivity	.32 to .35 per cent of standard annealed copper.
Thermal conductivity	Approx. same, relative to copper, as electrical conductivity.

28. This phase of the casting development has not been thoroughly investigated, but quite a number of tests have been made by two or three companies. Table 1 gives a few of these tests for alloys of different percentages, in "as cast," annealed and precipitation hardened states. All tests were made by the Chase Copper & Brass Co., Waterbury, Conn., the Bridgeport Brass Co., Bridgeport, Conn., and the American Manganese Bronze Co., Holmesburg, Pa. Variations in heat treatment and in contained beryllium will alter these results.

Table 1
PHYSICAL TESTS OF BERYLLIUM-COPPER ALLOYS

Per Cent of Beryllium.	Elastic Limit, lbs. per sq. in.	Tensile Strength, lbs. per sq. in.	Elongation in 2 in., per cent.	Reduction of Area, per cent.	Brinell No.
"As Cast" Condition:					
1.79	47,130	60,200	5.0	11.6	119
1.79	38,200	63,400	9.0	13.1	100
1.36	33,400	41,500	4.0	9.7	136
2.19	43,600	62,200	14.0	20.0	109
2.19	39,100	47,100	7.0	13.0	93
2.19	51,800	72,200	10.0	22.0	158
Annealed Condition:					
1.67	16,810	41,100	29.0	31.4	54
1.79	15,800	33,400	16.0	22.2	65
2.19	22,800	39,250	11.0	22.0	86
2.19	22,150	40,600	13.0	25.0	70
2.45	28,400	57,200	6.0	13.0	143
Precipitation Hardened at 575 degs. Fahr. for 3 hrs.:					
1.79	39,000	71,100	3.0	5.5	119
2.19	84,800	104,000	1.0	0	444
2.19	87,600	116,600	1.0	0	415
2.45	93,400	117,300	1.0	0	415
2.45	97,700	117,000	1.0	0	415
Precipitation Hardened at 575 degs Fahr. for 24 hrs.:					
1.67	49,300	69,600	3.0	3.1	183
1.79	74,200	95,400	2.0	1.2	207
2.50	83,200	100,000	1.0	0	269
1.57	73,600	91,300	2.0	0	321
1.57*	74,200	99,600	3.0	0	331

* Heated for 7 hrs.

Table 2
PHYSICAL PROPERTIES OF ROLLED BERYLLIUM-COPPER ALLOYS AFTER HARDENING

	Johnson's Elastic Limit, lbs. per sq. in.	Tensile Strength, lbs. per sq. in.	Elongation in 2 in., per cent.	Rockwell Hardness No.	Young's Modulus, $\times 10^3$
1.5 per cent Beryllium:					
Fully annealed	—	49,000	64.5	B25	—
Rolled 6 Nos. hard....	—	99,000	4.0	B96	—
Same, Heat-treated (Precipitation Hardened)	—	128,000	11.0	B104	—
2.0 per cent Beryllium:					
Fully annealed.....	15,000	59,000	65.0	B42.5	15.2
Rolled 6 Nos. hard....	79,000	112,000	4.0	B100	16.0
Same, Heat-treated (Precipitation Hardened)	132,000	176,000	2.5	B113	17.9
2.5 per cent Beryllium:					
Rolled 1½ Nos. hard...	64,000	110,000	10.0	C22	16.3
Same, Heat-treated (Precipitation Hardened)	116,000	196,000	3.0	C42	19.0

Properties of the Rolled Alloys

29. The physical properties of these alloys are quite different after rolling and precipitation hardening. A few of such results are interjected here as a matter of general interest. For example, according to tests made by the American Brass Co., Waterbury, Conn., the properties of 0.040-in. sheet are given in Table 2. Variations in the results in the table are possible, depending on composition, thickness of sheet, heat treatment and so on.

Electrical Conductivity

30. Compared with steel, phosphor bronze and other high-strength materials, the electrical conductivity of beryllium-copper alloys is reported high. This has been determined by the researches of two American companies. The same heat treatment which improves the physical properties so remarkably, as discussed in preceding paragraphs, also increases the electrical conductivity, as shown in Fig. 1. In general, this shows that the highest conductivity is reached at a beryllium percentage of around 2 per cent and a precipitation hardening period of 48 hours.

31. In further reference to the deoxidizing effect of beryllium on copper castings, it has been ascertained that whereas, for example, 0.02 per cent phosphorus introduced into copper castings results in an electrical conductivity of 44, only 0.01 per cent beryllium imparts an electrical conductivity of 53. Conductivities as

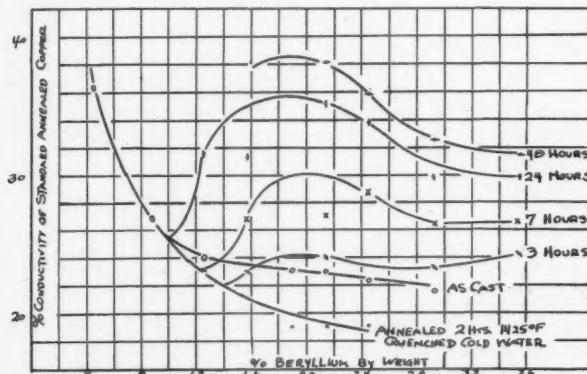


FIG. 1—ELECTRICAL CONDUCTIVITY OF BERYLLIUM-COPPER ALLOYS IN "AS CAST" CONDITION, SHOWING CHANGE IN CONDUCTIVITY WITH INCREASED TIME OF PRECIPITATION HARDENING TREATMENT. HARDENING FURNACE TEMPERATURE OF 575 DEGS. FAHR. \pm 15 DEGREES (BERYLLIUM DEVELOPMENT CORP.)

high as 56 are obtainable by the use of 0.02 to 0.03 per cent beryllium. It is claimed also that, in a similar manner, the heat conductivity of copper is also increased. The foregoing figures* are based on a conductivity of pure copper of 60 at 20 degs. Cent. (68 degs. Fahr.)

Thermal Conductivity

32. Some thermal conductivity tests have been made by a reliable laboratory, but further data are necessary for drawing satisfactory conclusions. The tests were made with two cylindrical specimens machined from a cast bar containing 2.45 per cent beryllium. The specimens were tested (a) in the as cast or "as received" condition, (b) in the quenched condition, (c) in the quenched and hardened condition, and (d) in the quenched and annealed condition.

33. In all three quenching treatments the specimens were heated to 1450 degs. Fahr., held for 1 hour and quenched in cold water. The specimens were hardened at 570 degs. Fahr. and air cooled. Forty-five minutes were required to reach this temperature, which was maintained for three hours before cooling. The specimens were annealed by heating at 750 degs. Fahr. for 50 hours, followed by a 20-hour cooling period in the furnace. For condition (c) the thermal conductivity in calories per second per centimeter per degree ($\text{sec}^{-1}\text{cm}^{-1}\text{deg}$) is as follows:

	Degs. Cent.								
	50	100	150	200	250	300	350	400	450
	Degs. Fahr.								
Quenched and hardened	122	212	302	392	482	572	662	752	842
	... 0.25	0.27	0.30	0.33	0.34	0.41	0.47	0.46	0.43

34. The drastic effect of temperature on the thermal conductivity of the 2.50 per cent alloy of beryllium-copper is quite unusual and interesting, says the report, as a means for studying structural changes in age-hardening alloys. No doubt the sharp increase in conductivity is related in some way to precipitation taking place within the alloy; but in view of the unusual nature of the tests, an attempt to interpret the results is not justified until more information is available.

Hardness Values

35. Some hardness tests of cast 1-in. rods of two different beryllium-copper alloys have been made recently by the Martin

* Data taken from "Beryllium-Copper Alloys," E. K. Thews, Berlin, Germany. *Canadian Chemistry and Metallurgy*, March 1931.

Table 3
HEAT TREATING EFFECT ON HARDNESS* OF TWO ALLOYS

Specimen.	Time, hrs.	525 degs. Fahr.		575 degs. Fahr.	
		Beryllium, 1.85%	Beryllium, 2.35%	Beryllium, 1.85%	Beryllium, 2.35%
A	1	112	312	107	269
		B72	B109	B67	B102
B	1½	143	269	137	340
		B75	B106	B78	B112
C	2	207	302	156	293
		B98	B107	B104	B110
D	2½	241	302	302	375
		B100	B109	B107	B112
E	3	269	351	302	302
		B102	B111	B107	B114

* Figures for Rockwell hardness indicated by B; Brinell hardnesses have no indicating letter.

Forge Co., Brooklyn, N. Y., which are of some interest. The two alloys contained 1.85 per cent and 2.35 per cent beryllium, respectively. They were heat treated in molten salts at 525 and 575 degs. Fahr., respectively, for various periods as indicated in Table 3, which gives details of the resulting hardness in Brinell and Rockwell numbers.

36. These results, which are not conclusive but are indicative of the trend, show that at 525 degs. Fahr. the maximum hardness is reached at about the 3-hour heating period for both analyses. For the 575-degree heating period, the highest hardness seems to be reached at about the 2½-hour heating period. Variations in the progressive increases in hardness may be due to alterations in structure. This phase of the entire subject merits careful study.

High Wear-Resistance Properties

37. Some authoritative preliminary tests of the wear-resisting qualities of cast beryllium-copper alloys have been made. While the tests are by no means exhaustive, they substantiate in a measure the expectations which the precipitation hardening properties would warrant.

38. The wear tests were made on an Amsler machine and conducted according to the latest approved practice. As a standard for comparison, a 0.43 per cent carbon steel was used, oil quenched from 1600 degs. Fahr. and tempered at 1100 degs. Fahr. The resulting hardness was about C22 Rockwell. Cast disks of the alloy, containing 2.35 to 2.46 per cent beryllium, in six disks were

used. The tests were made to determine the effects of age or precipitation hardening on the wear-resistance properties of the alloys.

39. The hardness of the unheated or "as cast" disks was B73.1 to B85 on the Rockwell machine. After heating to 1479 degs. Fahr., holding for $\frac{1}{2}$ hour, quenching in water, then heating to 600 degs. Fahr. and holding for 3 hours, followed by air cooling, the hardness of two specimens was Rockwell C41.3 and C43.3. Heating to 575 degs. Fahr. and holding for 24 hours, followed by normalizing, resulted in a lower hardness of C31.4 and C30.1.

40. Results of tests indicated that the beryllium bronze as cast, with a Rockwell hardness of B85, lost 0.0084 grams per 10,000 revolutions in the Amsler machine. The steel specimen had a wear loss of 0.0168 grams. After the 3-hour precipitation hardening treatment, the disk having a hardness of Rockwell C43.3 lost only 0.0003 grams, as compared with 0.0001 grams for the standard steel. Specimens subjected to the 24-hour precipitation hardening treatment, which naturally had a lower Rockwell hardness, showed considerably larger loss from wear.

APPLICATIONS

41. The development of beryllium-copper castings is so comparatively recent that applications for them have not been extensively explored. Doubtless, because of the many possible properties which can be developed by judicious heat treatment, many uses will come to light as the industry develops. Several however, have been found which are of interest. A few of these are briefly discussed in the following paragraphs.

Bearing Sleeves for Airplane Propellers.

42. In the aviation field an extremely interesting application of a bearing sleeve of 2.50 per cent beryllium-copper casting has been approved and adopted. New adjustable pitch propellers for airplanes have been perfected and are now being introduced which promise to radically change flying methods and hazards. The bearing sleeve referred to is an essential factor, and it is stated that the composition referred to is the only one that assures success.

43. As described by an authority, these adjustable pitch propellers require an intermediate sleeve between the shaft and the blade in order to provide a satisfactory bearing base to take care of the relative motion between the two parts. No small amount of research was necessary to develop a nonferrous material sufficiently

hard as to carry the motion without sticking or scuffing. The problem was solved by the use of cast beryllium-copper. In the hardened condition, this alloy is extremely hard, providing at the same time a metal which possesses excellent bearing characteristics.

44. These sleeve castings (Fig. 2) are made by the American Manganese Bronze Co. While the usual practice is to machine or process cast beryllium-copper in the annealed or quenched (soft) state, applying the hardening operation to the finished piece, study of the manufacturing process in this case has shown that it is cheaper and more satisfactory to complete the heat treatments first, doing all the machining after the final hardening heat.



FIG. 2—GROUP OF BERYLLIUM-COPPER SLEEVE BEARING CASTINGS FOR USE IN VARIABLE PITCH AIRPLANE PROPELLERS.

45. This fact seems to indicate that there is no difficulty, as a manufacturing process, in machining beryllium-copper in its finished condition when having a Brinell hardness of about 400. This use of cast sleeves points to the possibility of important applications for bearings where a hard, long-wearing nonferrous metal is desired.

Racing Boat Propellers of Beryllium Castings.

46. Another use of beryllium bronze, which has very recently been demonstrated, is for cast propellers of racing boats. A 3-blade propeller of 2.5 per cent beryllium-copper, weighing about 18 lbs., recently was cast by the Columbian Bronze Co., Freeport, N. Y. One reason for trying this was that it was possible to de-

crease the thickness 25 per cent from the manganese propeller which was displaced. There was no consequent loss in strength or rigidity. The propeller, after thorough tests, was found to increase the speed 2 seconds per half mile over any propeller previously used. It is being installed in one of the Gold Cup racing boats.

47. *New Kink in Heat Treatment.* A most interesting discovery was made in heat treating this propeller. Of course, to put it in its most efficient condition, it was necessary to give it the precipitation hardening treatment. When this had been done at 575 degs. Fahr. for 3 hours, it was found to be distorted or out of pitch. Strenuous efforts, using hammer blows on lead or steel blocks, failed to restore the original pitch—the casting sprang back after each blow to its distorted condition, indicating not only the rigidity but also the resistance to impact and shock of this alloy.

48. Experiments in heat treatment finally solved the difficulty. The propeller was put through a modified precipitation heat treatment. It was first heated to 500 degs. Fahr. for 1 hour and normalized; then this was repeated at 525 degs. Fahr. for 2 hours. The result was a propeller normal in pitch and in properties. This is believed to demonstrate the flexibility of the heat-treating practice when applied to this material.

Cams of Beryllium Copper.

49. It has been found that cams of beryllium bronze have some advantageous properties when installed in certain machines. A few of these have been cast having a diameter of 12 inches with a 7-in. face, and a large can manufacturer has been using them with success. It was found that in certain but not all of the large body machines, the cast iron cams were causing much trouble from wear and other causes. In the machine where the greatest trouble was experienced, two cams of 1.75 per cent beryllium-copper were substituted for the iron. The life of the latter cams had been very short, but in the new installation the cams have lasted over nine months and there has been no trouble in operating the machines.

50. For gears and pinions, particularly worm gears, this alloy is claimed to be particularly suitable. Some gears have been installed on motor buses for a certain company, but data on the results are not yet available. It is expected, from the experience thus far, that they will prove efficient and satisfactory.

As a Corrosion-Resistant Material.

51. In the field of corrosion resistance, several applications of fabricated beryllium-copper alloys indicate promising possibilities for castings in the petroleum industry. In two installations, handling extremely corrosive and abrasive sulphur compounds, the performance of the alloy is materially superior to any of the many alloys previously tried.

52. Some other miscellaneous applications which are under way include worm driving gears for jaw riveting machines, chilled rifle bar nuts for rock drills, propeller hub cones for airplanes, and valve parts and centrifugal pump parts for line valves.

General Applications.

53. There is also available another beryllium master alloy of nickel and copper with which some interesting products are anticipated. One suggested application for this alloy is its introduction into nickel alloy castings such as Everbrite and monel metal. A large manufacturer of castings of these alloys states that there is a constant demand for a harder material of approximately these compositions. Experiments are under way to add beryllium to these castings to meet this and other properties. Also, from the deoxidizing standpoint, one maker of machinery has found that the use of beryllium as a deoxidizing agent in high-nickel castings has resulted in a decided improvement in the soundness and quality of his castings.

54. The whole field of beryllium alloy nonferrous castings has been hardly scratched. It is not improbable that small additions of other alloys to the beryllium-copper castings may be necessary to overcome microstructural irregularities, to improve properties, and to make them meet other conditions.

55. In conclusion, it should be repeated that this presentation of the subject is merely in the nature of a progress report of a new development from which much is expected as research metallurgists delve into its many possible phases, both explored and unexplored.

DISCUSSION

CHAIRMAN SAM TOUR¹: I might add a remark or two on the subject of beryllium as a deoxidizer. In paragraph 18 on page 334, the statement is made that standard practice is to add 0.01 to 0.02 per cent beryllium, either alone or, preferably, following standard deoxidizers such as boron.

¹ Lucius Pitkin, Inc., New York.

copper. I might add a note of warning that phosphorus, which is used commonly as a deoxidizer of copper, is harmful in beryllium-copper alloys. The presence of phosphorus tends to interfere with the precipitation hardening properties of beryllium and copper. Copper should not be deoxidized with phosphorus if beryllium is added to it.

In addition to boronic copper for deoxidization, calcium, silicon, lithium, magnesium, or combinations of those elements can be used satisfactorily.

C. S. COLE²: Mr. Riley in presenting this paper said the modulus of elasticity was rather high. What was the order of the modulus?

W. L. RILEY³: Eighteen to 18½ million lbs. per sq. in. in a 2½ per cent alloy.

CHAIRMAN TOUR: Mr. Riley, can you record anything on the difficulties encountered if you get beryllium contamination in lead and tin bronzes?

W. L. RILEY: We have not heard yet of any reaction on the appearance of beryllium in scrap metals. The casting of beryllium copper alloys has been confined to a comparatively limited number of foundries. In both the fabricated and the cast condition, there is a considerable amount of beryllium alloy in use now and eventually that problem will require solution.

Some work has been done in the addition of small fractions of a per cent of beryllium to the 80-10-10 or the other series of bronzes. As far as the effect on the physical properties is concerned, nothing beneficial has been found. As a matter of fact, there may be detrimental effects and, consequently, it possibly may be expected that the appearance of beryllium in scrap will be something to watch out for and to guard against. Beryllium, with its tremendous affinity for oxygen, which is the difficulty that held up the development of reduction processes so long, will tend to dispose of itself in very small fractions.

Although the protection of beryllium from oxidation in the reduction process has been such a difficult matter, in foundry practice where there are suitable and capable metallurgical facilities and care is taken, it has not been difficult to learn the technique of beryllium-copper casting, with confinement of the loss to an extremely moderate amount of contained beryllium.

T. C. WATTS⁴: We were asked to melt some copper for a local concern and to use some beryllium. We cast two bars one inch in diameter and 12 inches long. While I did not get the full reports, the customer's metallurgist told me that the conductivity was low. As I remember it, he said that the conductivity was approximately 7 per cent on the bar as cast and after heat treating it increased slightly. He was disappointed in the alloy as a conductor. However, he said that he had a very high grade chisel. I think that it could be used on steel and iron, it was so hard.

² Copper and Brass Research Assn., New York.

³ Beryllium Products Corp., New York.

⁴ Falcon Bronze Co., Youngstown, Ohio.

CHAIRMAN TOUR: Conductivity of copper is affected materially by very slight traces of iron. The average foundry practice of stirring copper with an iron bar is sufficient to introduce enough iron to completely ruin the conductivity. If the conductivity was as low as 7 per cent, I hardly would say that it should be blamed on beryllium because the fact is that with as high as 1.5 to 2 per cent of beryllium, close to 20 per cent conductivity is secured easily.

T. C. WATTS: I do not know about their method or where they got their conductivity values but we melted the copper in a new crucible, stirred it with a graphite rod, and it was handled under as nearly perfect conditions as possible.

MEMBER: Foreign practice indicates great difficulty in re-using beryllium-copper scrap due to the formation of blowholes in the castings. A number of European companies have discontinued the use of beryllium-copper for that reason.

CHAIRMAN TOUR: I think possibly that objection is slightly unfounded, based upon actual experience in this country insofar as remelting beryllium-copper as such is concerned. There have been no particular difficulties in remelting and reusing beryllium-copper alloys as beryllium-copper. I know of one case where a foundry making beryllium-copper castings mixed beryllium-copper scrap in the form of gates and sprues with some bearing metals. It has given them trouble due to contamination in other types of alloys. However, there has been no particular difficulty in this country in the remelting of beryllium-copper scrap.

W. L. RILEY: I think what Mr. Tour said would be borne out by reference to those companies using beryllium-copper in this country. Those foundries producing beryllium-copper castings regularly use their scrap. In using scrap, I think perhaps what Mr. Tour refers to, originates in the practice where foundries attempt to melt the scrap. The only way in which beryllium-copper scrap should be used is as an addition to already melted copper, with sufficient of the master alloy to balance the percentage to whatever is desired in the final alloy. There are cases where attempts have been made to re-melt scrap direct. After the material had been in the bottom of the crucible for some time, the melter was surprised to find that it was still solid, due to the formation of such a perfect coating of beryllium oxide on the outside. While the interior may have been plastic or possibly molten, the scrap did not change its shape. It is bad practice to expose beryllium-copper scrap or the master alloy to the furnace temperature because of the oxidation of the beryllium. It can satisfactorily be used only as an additive.

The Mechanism of Inverse Segregation*

By OWEN W. ELLIS,† TORONTO, CANADA

I—Introduction

1. Our purpose is to discuss certain defects which are encountered in bronze foundry alloys of the 89-11 type and, in particular, such defects as apparently result from the presence within these alloys of gaseous or solid non-metallic elements or compounds.

2. By way of introducing the subject for discussion, the author will take the liberty of describing rather closely some of the chemical and physical characteristics of the alloys with which we are to deal, since this may promote the objects for which we are met—the free interchange of ideas on the deoxidation and degasification of tin-bronzes.

3. The alloys to be dealt with in our symposium are composed essentially of copper and tin, to which phosphorus, manganese, zinc or other elements may be added for the purposes of deoxidizing the alloy only, or of improving its mechanical properties also, *e. g.*, by adding one or more of these elements in amounts exceeding those required for purposes of deoxidation only. Dews,¹ for example, has shown that phosphorus up to about 0.05 per cent has little or no effect on the mechanical properties or structure of government bronze or Admiralty gunmetal. The amount of tin in these alloys varies from about 8 to 12 per cent in good practice. It rarely, if ever, falls below or rises above these values.

4. Some reference should be made to lead, which can be looked upon neither as a deoxidizer nor, for that matter, as an

* An introduction to the Symposium on Deoxidation and Degasification of Bronze Foundry Alloys, held June 22, 1933, during the A.F.A. Annual Convention. This material refers particularly to the third alloy group (Tin-Bronzes, 89-11 type), discussed at the symposium. Discussion will be found on page 377.

† Director of Metallurgical Research, Ontario Research Foundation.

¹ *Journal, Institute of Metals*, 1930, v. 44, p. 255.

element entirely deleterious in its action on these alloys. Rolfe's investigations² and the discussion provoked by their publication appear to have proved definitely that lead in the usual amounts—not more, for example, than the 3 per cent sometimes added to improve machinability—has a distinctly beneficial effect on the mechanical properties of alloys of this type. Both paper and discussion made it clear, on the one hand, that the manufacture of gunmetal containing up to 3 per cent lead was unattended by difficulties exceeding those ordinarily encountered in bronze foundry practice, and, on the other hand, that both the tensile strength and the ductility of the alloy were improved by additions of lead up

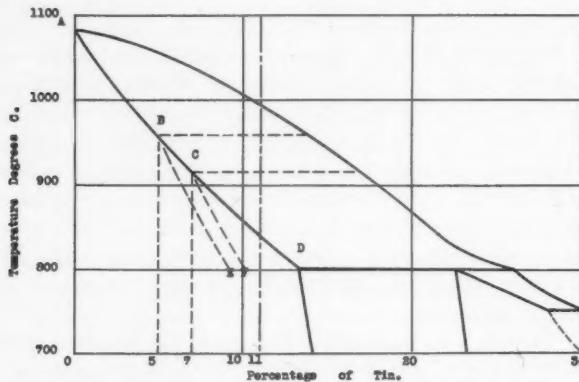


FIG. 1—EQUILIBRIUM DIAGRAM OF THE COPPER-TIN SYSTEM.

to 1.5 per cent; even the resistance of the alloy to impact (Stanton test) was unimpaired by the presence of as much as 1 per cent of the soft metal.

II—RAPID COOLING AND ITS EFFECT ON THE DISTRIBUTION OF TIN IN 89-11 BRONZE

5. The equilibrium liquidus (see Fig. 1) of the 89-11 alloy of the copper-tin system, the alloy typical of those under discussion, is in the neighborhood of 1819 degs. Fahr. (993 degs. Cent.); the equilibrium solidus is close to 1539 degs. Fahr. (837 degs. Cent.) However, the rates of cooling ordinarily met with in the foundry are such that this alloy remains molten at temperatures well below the solidus—1539 degs. Fahr. (837 degs. Cent.)—which

²*Journal, Institute of Metals*, 1921, v. 26, p. 85.

the foundryman equipped with pyrometer can prove for himself by placing the hot junction of a couple in the gate or riser of a suitable casting. He will find that not until the metal has reached a temperature of between 1472 and 1418 degs. Fahr. (800 and 770 degs. Cent.) will its solidification prevent the removal of the couple from the casting.

6. Why is it, it may be asked, that an alloy such as this which, under conditions of equilibrium, becomes wholly solid at 1539 degs. Fahr. (837 degs. Cent.) should remain partially liquid until a temperature somewhere between 1472 and 1418 degs. Fahr. (800 and 770 degs. Cent.) has been reached? The fundamental reason is that the alloy, when poured into molds of the dimensions usual in bronze foundry practice, cools at such a rate that primary crystallization is delayed until the metal in the mold has in many instances reached a temperature as low as 1679 degs. Fahr. (915 degs. Cent.)—a point half way between the equilibrium liquidus and solidus. There is pyrometric evidence to support this view, which is further sustained by such studies as have been made of the microstructure of sand and chill castings of alloys of this type cooled at different rates.

7. If spontaneous crystallization is delayed until a temperature of 1679 degs. Fahr. (915 degs. Cent.) has been reached, the crystal nuclei which first form in the melt at or near the surface of the mold will contain approximately 7 per cent tin. Upon these nuclei there will build up the peripheral dendrites normal to the mold surface, which are characteristic of all castings, whether sand or chill. The solid at the surfaces of these dendrites in immediate contact with the melt will at any given temperature coincide in composition with that denoted by the points on the equilibrium solidus (line *ABCD*, Fig. 1) corresponding with these temperatures.

8. While some diffusion of tin within the growing peripheral dendrites is certain to occur, it would appear that the movement of tin inward from the surfaces of these crystallites is relatively slow at temperatures between 1679 and 1472 degs. Fahr. (915 and 800 degs. Cent.), and that the nuclei are enriched but little in tin during their rapid cooling through this range. If no diffusion occurred in the dendrites, their mean composition at any temperature would be represented by a point on the line *CF*, midway between the 7 per cent and 13.3 per cent ordinates of the equilibrium diagram. Were these conditions fulfilled (and, apparently, they

are not so very different from those fulfilled in small castings), the mean composition of the peripheral dendrites, on reaching the peritectic temperature of 1472 degs. Fahr. (800 degs. Cent.) would be in the neighborhood of 10 per cent tin, while the tin content of the melt would tend toward 29 per cent. Such few globules of melt as were so enmeshed in the interlacing branches of the peripheral dendrites as to be prevented from freely mixing with the body of liquid metal in the center of the casting might actually attain to this high content of tin. However, although the tin is likely to be more concentrated in the melt in the immediate neighborhood of the peripheral dendrites, the gradient of concentration in the melt at the center of the casting is likely to be low, that is, the difference in the tin content of the melt at points fairly near to, and remote from, the dendrites is likely to be small. Nevertheless, the melt at the center of the casting may be considered as abnormally, even though slightly, enriched in tin.³

9. This abnormal enrichment of the melt in tin might be expected to have an effect upon the composition of those nuclei which sooner or later will crystallize spontaneously and promiscuously throughout the melt in the body of the casting. Such crystallization will not necessarily be deferred until the peripheral dendrites have ceased to grow, and will rarely if ever be delayed until the melt in the immediate vicinity of the peripheral dendrites, or such as has been enmeshed in the branches of these crystallites, has solidified. Nor will the temperature at which it occurs coincide, of necessity, with that at which the peripheral nuclei formed, the chances being that the melt in the body of the casting will not be under-cooled to the degree that those portions of the melt which first make contact with the mold are under-cooled. Hence, although the abnormal enrichment of the melt in tin, due to the formation of the peripheral dendrites, might be expected to lead to the precipitation of nuclei richer in tin than those which precipitated at the periphery of the casting, because spontaneous crystallization within the enriched melt occurs at higher temperatures than 1679 degs. Fahr. (915 degs. Cent.), the effect of the enrichment of the melt on the composition of the internal nuclei will be set aside and these nuclei, when they form, will most probably contain less than 7 per cent tin.

³The normal enrichment of the melt is that which occurs when the rate of cooling is such that the dendrites, whether peripheral or internal, are uniform in composition throughout at every temperature during cooling and become completely solid at the solidus, *i. e.*, the temperature at which the alloys of the system become completely solid under conditions of equilibrium.

10. If, for example, the spontaneous and promiscuous appearance of nuclei in the melt within the body of the casting were to occur at 1760 degs. Fahr. (960 degs. Cent.), the tin content of the nuclei would be in the neighborhood of 5 per cent; and, were no diffusion to occur within the dendrites as they formed round these nuclei during cooling from 1760 degs. Fahr. (960 degs. Cent.) to the peritectic temperature of 1472 degs. Fahr. (800 degs. Cent.), the mean tin content of these internal dendrites on reaching this temperature would be in the neighborhood of 9 per cent—the melt in contact with the dendrites would contain, at the same temperature, about 30 per cent tin.

11. Since the rates of cooling of the central parts of the casting will almost inevitably be less than those of the outer portions, the chances are that, owing to some diffusion of tin from the surfaces of the dendrites toward the nuclei, the concentration gradient of tin in the internal dendrites will be less than that in the peripheral dendrites. The slower the rate of growth of the internal dendrites, the richer in tin (mean composition) they will be, and the poorer in tin will be the portions of the melt last to solidify.

12. The tin enrichment of the melt in the central portions of the castings (due to the formation of the peripheral dendrites) would always express itself, of course, as an enrichment of the solid material at the inside of the casting, were it not that other factors occasionally enter into play which tend to raise the tin content of the outer parts of the casting (inverse segregation or liquation). These factors increase in importance as the rates of cooling of these alloys are increased, whether by the use of suitable pouring temperatures or by casting into chill molds.

III—INVERSE SEGREGATION

13. One explanation⁴ for inverse segregation or liquation is that the hydrostatic pressure exerted by the contracting shell of peripheral dendrites on the mushy interior of the casting leads to inter-dendritic extrusion through the weak walls of the casting of portions of the tin-rich melt, although it seems hard to believe that the coefficient of contraction of the solid shell could under any circumstances exceed that of the mushy interior and thus bring about the observed effect. That gases coming out of solution in the alloy during solidification may force portions of tin-rich melt

⁴ Kühnel, *Zeit. Metallkunde*, 1922, v. 14, p. 426; and others.

toward, if not through, the surface of the casting has been suggested by Genders,⁵ whose interesting theory is discussed more fully in paragraphs 56 to 60, inclusive.

14. Both the phenomenon of under-cooling⁶ and the Ludwig-Soret effect⁷ have been used in the attempt to explain the mechanism of inverse segregation. In the author's opinion, however, the inter-dendritic flow theory of Bauer and Arndt⁸ fits the facts most closely. This theory "suggests that between the growing crystals cavities are formed as a result of contraction, and that molten alloy from the interior flows toward the surface to fill these cavities, becoming more concentrated during its passage by the rejection of solid of low concentration from its boundary. The liquid arriving at the surface layer thus takes the form of inter-dendritic fillings of high concentration. An alloy in which the primary dendrites have a low velocity of axial growth is assumed to give normal segregation, since liquid cannot extend between the crystals, and where diffusion is rapid the liquid of high concentration is carried toward the central region."⁹ In discussing this theory, Genders objects that: "The inter-dendritic channels along which the concentrated liquid must move are of very small cross-section and, although the viscosity of the liquid is probably low," (see paragraph 43) "the effect of friction may be considerable, so that it is difficult to regard it as possible for the passage, during the short time occupied in solidification, of concentrated liquid in appreciable quantity to the outer-most layer." This would be true, were it not that the liquid does not *meander* under the action of gravity toward the surfaces to fill the cavities formed as a result of contraction, but, given the requisite conditions, which the author describes below (in paragraphs 37 to 49, inclusive), is continuously *forced* into the cavities during the last stages of solidification.

IV—TIN SWEAT

15. As hinted above, the author has developed a theory which seems to account for the movement of tin-rich liquid to the periphery of rapidly cooled castings, but before dealing with this he would like to discuss one or two other points, beginning with an interesting photograph which illustrates Genders' paper, and which has been reproduced here (Fig. 2). This photograph might well

⁵ *Journal, Institute of Metals*, 1927, v. 37, p. 241.

⁶ Masing, *Zeit. Metallkunde*, 1922, v. 14, p. 204.

⁷ Benedicks, *Transactions A.I.M.E.*, 1925, v. 71, p. 595.

⁸ *Zeit. Metallkunde*, 1921, v. 13, p. 559.

⁹ Genders, *Journal, Institute of Metals*, 1927, v. 37, p. 241.

be that of two gates which the author himself investigated shortly before the publication of Genders' paper. However, whereas the gates shown in Genders' photograph were the tops of vertical bars of phosphor-bronze (89-10-1 type), those in which the author was interested had been cut from sand-cast bearings of the 80-10-10 type.

16. Genders describes the globule which characterized the top of one of his phosphor-bronze bars as "of high-tin alloy," but gives no analysis. It probably contained at least 17 per cent tin. The exudation on the gate of the sand-cast bearing referred to above contained over 20 per cent tin, the lead content of the

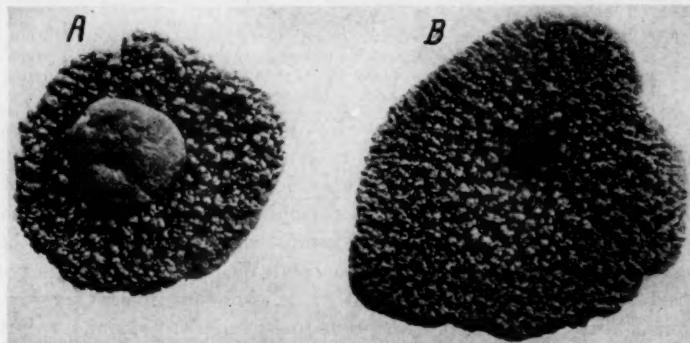


FIG. 2—A: TOP OF SAND-CAST BAR OF PHOSPHOR-BRONZE MELTED AND CAST NORMALLY. APPROXIMATELY ACTUAL SIZE. B: TOP OF BAR FROM SAME MELT AS A; CAST AFTER SOLIDIFICATION IN CRUCIBLE AND REMELTING IN ELECTRIC FURNACE. REDUCED APPROXIMATELY ONE-THIRD IN PRINTING. (R. GENDERS)

button, however, being practically the same as that of the body of the casting. The button was extruded from the gate when the metal in its immediate vicinity was in process of solidification—at a temperature of about 1436 degs. Fahr. (780 degs. Cent.). Clamer¹⁰ has referred to similar exudations on the surface of phosphor bronze poured into open ingot molds: "An analysis of these nodules would indicate a tin content varying from 13 to as high as 22 per cent, whereas the body of the pig contains only 10 per cent tin. The phosphorus content in these nodules might be increased from probably 0.6 or 0.7 in the pig up to as high as 3 per cent in the nodules."¹¹

17. Neither Genders, Clamer nor Dews refers to the macro-

¹⁰ *Proceedings, A.I.M.E.*, Institute of Metals Div., 1927, p. 89.

¹¹ See also "The Metallurgy of Bronze," Dews. *Sir Isaac Pitman & Sons, Ltd.*, 1930, pp. 134-137.

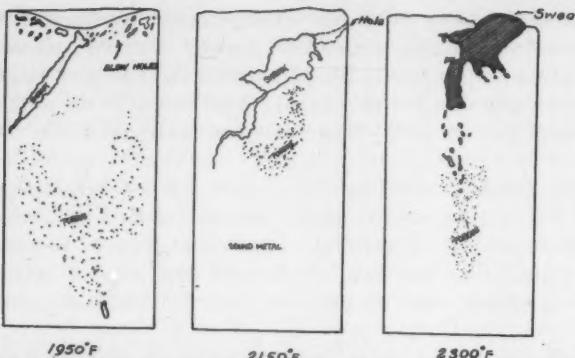
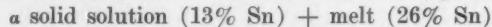


FIG. 3—EFFECT OF POURING TEMPERATURE, SHOWING CAULIFLOWER HEAD. TOP POURED; NOT OPEN TO AIR; GREEN SAND. ALLOY, TIN 8 PER CENT, LEAD 2 PER CENT, ZINC 4 PER CENT, COPPER 88 PER CENT. BARS 4 INCH DIAMETER, 10 INCH LONG. (BOLTON AND WIEGAND)

structure of the castings characterized by these high-tin exudations. The author believes that a comparison of polished and, preferably, etched sections of these with others devoid of exudations would have revealed in the former, macroscopic porosity of the type so well described by Bolton and Weigand,¹² whereas the latter would have been relatively sound. Bolton and Weigand's Fig. 11 is reproduced here as Fig. 3. The author believes that inverse segregation is always coincident with pronounced incipient shrinkage, but that incipient shrinkage is not of necessity accompanied by inverse segregation.

18. All these investigations support the view that the reaction between solid solution and melt which may be presumed to occur at the peritectic temperature under conditions of equilibrium, *viz.*,



will rarely if ever occur under the usual conditions of foundry practice, and particularly in the case of chill castings and small castings. In brief, when on cooling from above 1832 degs. Fahr. (1000 degs. Cent.), 89-11 bronze approaches the peritectic temperature, it comprises solid (dendrites) containing from $9 \pm$ to $10 \pm$ per cent of tin (mean composition of dendrites) and liquid containing from $29 \pm$ to $30 \pm$ per cent tin. The reaction between the solid and liquid phases at the peritectic temperature is so slight as to be almost negligible. Hence, the liquid transforms into β solid solution containing from 29 to 30 per cent tin and does

¹² "Incipient Shrinkage in Some Nonferrous Alloys," *Transactions A.I.M.E.*, Institute of Metals Div., 1929, p. 475.

not solidify completely until temperatures in the neighborhood of 1382 degs. Fahr. (750 degs. Cent.) are reached. The β solid solution changes into a β' - δ complex at 1094 degs. Fahr. (590 degs. Cent.), into an α - δ complex at 968 degs. Fahr. (520 degs. Cent.), but undergoes no appreciable alteration in structure on cooling from 968 degs. Fahr. (520 degs. Cent.) to room temperature.

V—EFFECT OF POURING TEMPERATURE AND RATE OF COOLING ON THE DENSITY OF 89-11 BRONZE

19. The density of the two constituents in this alloy is well worthy of consideration in this connection. There seems but little doubt that those portions of the alloy which are last to solidify—the tin-rich portions which transform into the α - δ complex—have a density¹³ in the close vicinity of 8.92; experiment has shown that bronzes containing from 30 to 31 per cent tin have practically the same density (8.92), whether they be cast in sand or in chill. This need scarcely be a matter for surprise, seeing that the freezing range of these high-tin alloys is less than 36 degs. Fahr. (20 degs. Cent.) and that there are few complexities in their process of solidification.

20. In view of the fact that the α - δ complex is of almost invariable density, any variations in the density of the 89-11 bronze must be due to one or other of a number of causes, of which the following are probably the most important:

(a) The high-tin liquid, when it solidifies, shrinks away from the dendrites, leaving microscopic cavities (see paragraph 39)—such, for example, as have been described by Bolton and Weigand in their papers on "Incipient Shrinkage in Some Nonferrous Alloys"¹⁴ and "Effect of Oxidation and Certain Impurities in Bronze."¹⁵

(b) True blowholes form in the castings, owing to the release of gases from solution.

21. Now, of all the industrial alloys of the copper-tin system, the 89-11 is most subject to variations in density. Bailey,¹⁶ for example, found the density of 1 $\frac{1}{4}$ -in. diameter bars (10.9% Sn, 0.53% P) to vary from 7.61 to 8.31 as the pouring temperature was lowered from 2192 to 1850 degs. Fahr. (1200 to 1010 degs. Cent.): "With descending casting temperature, the density rose sharply to 8.18 and then gradually to 8.31. The best all-around

¹³ Heape, *Journal, Institute of Metals*, 1923, v. 29, p. 467.

¹⁴ *Transactions A.I.M.E.*, *Institute of Metals Div.*, 1929, p. 475.

¹⁵ *Transactions, A.I.M.E.*, *Institute of Metals Div.*, 1930, p. 368.

¹⁶ *Journal, Institute of Metals*, 1923, v. 30, p. 401.

mechanical properties were obtained in the bar poured at 2102 degs. Fahr. (1150 degs. Cent.), which had a density of 8.18. Tests made on a series of bars of lower phosphorus content (10.6% Sn, 0.15% P) showed that the density rose only gradually to 8.20 in the bar cast at 2147 degs. Fahr. (1175 degs. Cent.) and then very gradually to 8.29 as the casting temperature fell to 1904 degs. Fahr. (1040 degs. Cent.) Casting at 1832 degs. Fahr. (1000 degs. Cent.) resulted in a sudden increase in density to 8.50. Mechanical properties showed a maximum in the bar cast at 2147 degs. Fahr. (1175 degs. Cent.) and fell off quickly and uniformly as the casting temperature fell to 1904 degs. Fahr. (1040 degs. Cent.). In the bar cast at 1832 degs. Fahr. (1000 degs. Cent.) a sharp improvement in all properties was again found."

22. As a result of his investigations, Bailey concluded that (a) the slower the rate of cooling of a casting of this alloy, the lower is its density, (b) a considerable rise in density follows very rapid cooling, whether brought about by casting very thin bars or by pouring at a low temperature, and (c) very high density is also obtained by allowing the alloy to solidify in the pot.

23. Carpenter and Elam¹⁷ have suggested that "there are two distinct varieties of unsoundness" in government bronze or Admiralty gunmetal (88-10-2). In the first place, "when the metal is poured too hot . . . the metal actually rises in the mold . . . due to the formation of blowholes." In the second place, "when poured too cold . . . the surface sinks. Between the two extremes there is a temperature range in which sound castings are obtained, the surfaces of which are flat. When poured too hot, the metal is always unsound. The lower limit of unsoundness is not so well fixed, the casting generally showing holes, but not always." Confirmation of these findings has been given by Bolton and Weigand¹⁴ although their interpretation of the behavior of the alloy during cooling does not agree with that of Carpenter and Elam. It is natural to assume, however, that metal which "actually rises in the mold" should be of low density, and to this extent, therefore, Carpenter and Elam's observations confirm Bailey's conclusions regarding the effect of high pouring temperature on density.

24. As to the causes of blowholes: "their presence in gunmetal may be due either to slight dampness in the mold when the metal is cast into it, or the gases which form these holes may

¹⁷ *Journal, Institute of Metals*, 1918, v. 19, p. 155.

be actually dissolved in the metal when it is cast. . . . A microscopical examination is the surest way of ascertaining the cause of blowholes. If these are clean and confined to the outer portions of the castings, they are certain to be due to the steam formed within the mold, in which case a simple remelting brings the metal right. . . . More often the pin holes in castings appear dirty under the microscope, and on polishing, some brittle material is rubbed out, causing serious scratching of the surface. This is due to segregations of included oxides, either of tin or copper, which the small percentage of zinc has not been able to remove. Under these conditions remelting is of no avail unless with the aid of some deoxidizing agent such as phosphor-tin or phosphor-copper.¹⁸

25. The question of the internal appearance of blowholes in gunmetal has also been referred to by Parker:¹⁹ "While it is true that these were most frequently bright and free from colored oxide films, yet in actual castings they were somewhat tinged. It depended on whether air had gained access to them while still hot. This it sometimes did, owing to the presence of very fine intercrystalline cracks produced by shrinkage phenomena."

VI—WHAT CAUSES POROSITY IN 89-11 BRONZE?

26. It is open to considerable question, of course, whether the low density of all bronze castings poured at high temperatures can invariably be assigned to blowholes. Jenkin,²⁰ in discussing a paper by Rowe²¹ on the influence of mass and casting temperature on the physical properties of an alloy containing 87.54 per cent copper, 10.61 tin and 1.88 zinc, remarked: "that poor results" (i. e., of mechanical tests) "might be due to the presence of shrinkage cavities when the metal was poured at too high a temperature," and illustrated his point by photomicrographs of samples cut from castings of government bronze poured at 2462 degs. Fahr. (1350 degs. Cent.) This photomicrograph (Fig. 4) serves to confirm opinions which have been expressed by many versed in the art, that porosity in bronze castings poured at high temperatures is not always due to the occlusion of gas in the melt.

27. To sum up, while blowholes due to steam and, more rarely, to dissolved gases occur in bronze castings of the 89-11

¹⁸ Primrose, *Journal, Institute of Metals*, 1910, v. 4, p. 248.

¹⁹ Parker's discussion of Carpenter and Elam's paper in *Journal, Institute of Metals*, 1918, v. 19, p. 155.

²⁰ *Journal, Institute of Metals*, 1924, v. 32, p. 332.

²¹ *Journal, Institute of Metals*, 1924, v. 32, p. 327.

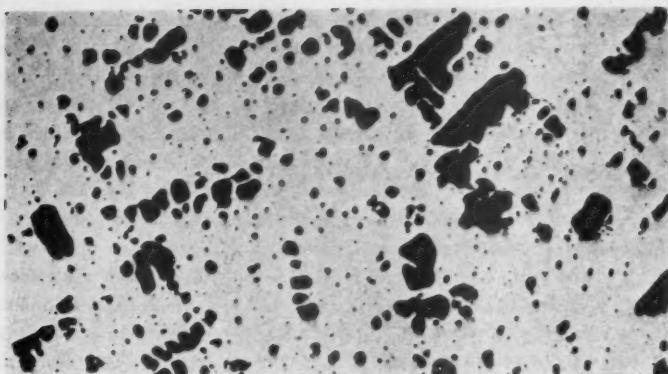


FIG. 4—ADMIRALTY GUNMETAL SAND-CAST AT 2462 DEGS. FAHR. (1350 DEGS. CENT.), SHOWING SHRINKAGE CAVITIES. X50. (J. W. JENKIN)

type, it is essential that they be distinguished from internal porosity of the type to which Jenkin referred, and which "surely is traceable to general incipient shrinkage," to use the term suggested by Bolton and Weigand in their earlier paper. "Shrinkage fissures and cavities are far too frequently confused with true 'blowholes' (gas holes)."²²

VII—EFFECT OF POURING TEMPERATURE AND COOLING RATE ON HARDNESS OF 89-11 BRONZE.

28. The increase in density which results from pouring bronzes of this type at decreasing temperatures seems to reflect itself in the results of hardness tests. It cannot yet be said that hardness and density go hand in hand, although Bailey's tests referred to in part above, and the results of Rowe's work²³ and of Dews²⁴ experiments, all tend to support this proposition. Rowe found the Brinell hardness numbers at the centers of $\frac{1}{2}$ -in., 1-in. and 2-in. square bars of government bronze to increase uniformly as the pouring temperature was depressed—from 2192 to 2012 degs. Fahr. (1200 to 1100 degs. Cent.)—and to vary inversely as the size of the bars. He further observed that the hardness at the centers of the 2-in. square bars was decidedly less than at the sides, the difference in hardness between the centers and the sides of the bars becoming less as the pouring temperature de-

²² Parker's discussion of Carpenter and Elam's paper in *Journal, Institute of Metals*, 1918, v. 19, p. 155.

²³ *Journal, Institute of Metals*, 1924, v. 32, p. 327; also *Journal, Institute of Metals*, 1924, v. 31, p. 217.

²⁴ *Journal, Institute of Metals*, 1930, v. 44, p. 255.

creased. The hardness at the sides of the bars was not so much affected by pouring temperature as that at the centers of the bars, save in the case of the bar poured at 2192 degs. Fahr. (1200 degs. Cent.), which, however, was only very slightly less in hardness at the sides than those poured at 2102 and 2012 degs. Fahr. (1150 and 1100 degs. Cent.), respectively.

29. It is worthy of note that Izod test pieces cut from the centers of the 1-in. bars were more resistant to impact than were similar pieces cut from the centers of $\frac{1}{2}$ -in. bars; and further, that Izod test pieces cut from the centers of the 2-in. bars were less resistant to impact than were pieces cut from the outsides of the same bars. A comparison (Table 1) of the samples cut from the 1-in. square bars with those cut from the outsides of the 2-in.

Table 1
Casting
Temperature.

	degs. Cent.	degs. Fahr.	Brinell Hardness.	Izod No., ft.-lbs.
1-in. square bar.....	1200	2192	57	11
2-in. square bar (outside).....	1200	2192	66	11*
1-in. square bar.....	1150	2102	61	10
2-in. square bar (outside).....	1150	2102	69 $\frac{1}{2}$	12*
1-in. square bar.....	1100	2012	65	10
2-in. square bar (outside).....	1100	2012	69	16*

* Mean of two test results.

Table 2
Casting
Temperature.

	degs. Cent.	degs. Fahr.	Brinell Hardness.	Izod No., ft.-lbs.
$\frac{1}{2}$ -in. square bar.....	1200	2192	61	5
2-in. square bar (center).....	1200	2192	49	3
$\frac{1}{2}$ -in. square bar.....	1150	2102	69	6
2-in. square bar (center).....	1150	2102	53	6
$\frac{1}{2}$ -in. square bar.....	1100	2012	80	6
2-in. square bar (center).....	1100	2012	56	4

square bars shows them to be much the same, the results on the whole favoring the 2-in. bars.

30. A comparison of the hardness and impact values of the samples cut from the $\frac{1}{2}$ -in. square bars with those of the samples cut from the centers of the 2-in. square bars shows that, while the impact values are much the same in both cases, the $\frac{1}{2}$ -in. bars in this case being favored, the hardness values of the $\frac{1}{2}$ -in. bars are distinctly higher than those of the 2-in. bars, as is shown in Table 2.

31. These results are most enlightening. To the author it seems clear that in the case of the $\frac{1}{2}$ -in. bars the coincidence of relatively high hardness and low impact number indicates relatively high density and the presence of the α - δ complex. The complex has contributed to the hardness of the material but, at the same time, has rendered it brittle. On the other hand, the coincidence of relatively low hardness and low Izod number in the samples cut from the centers of the 2-in. bars indicates relatively low density, and possibly a deficiency of the α - δ complex. The low resistance to impact is due in this case to the weakness of the structure and not to the brittleness of the α - δ complex.

32. If the author's interpretation of these valuable results is correct, the question arises: Why the relatively low density at the center of large castings?

VIII—EFFECT OF POURING TEMPERATURE AND COOLING RATE ON MICROSTRUCTURE OF 89-11 BRONZE

33. Here it is that the microscope may prove of use. Thirty-three years have passed since Longmuir²⁵ disclosed the important effect of pouring temperature on the microstructure of alloys of the type now under discussion. At a later date, Primrose²⁶ showed that: "With gunmetal castings the temperature of pouring has a great influence on the strength of finished castings. The chief causes for the differences which exist in the quality of the metal are the different rates of cooling and freezing. In the case of the pouring at high temperature, just before freezing commences the sand of the mold becomes slightly heated and so the freezing of the metal is slower than when the metal is not so hot at first; hence, the crystals have time to grow and form a good strong interlocking structure." The importance of this "interlocking design with uniformly intermixed primary crystallites and interspacial eutectoid" was emphasized again by Primrose²⁷ in 1913, although his opinion that such a structure would prevent porosity apparently underwent slight modification during the three years that elapsed between his two papers. During this period he seems to have found that, even when the interlocking

²⁵ Longmuir, "Brasses and Bronzes," *Sheffield Soc. of Engineers and Metallurgists*, March, 1900; also, Longmuir, "On the Influence of Varying Casting Temperature on the Properties of Alloys," *Journal, Iron and Steel Institute*, 1903, v. 63, p. 457.

²⁶ *Journal, Institute of Metals*, 1910, v. 4, p. 248.

²⁷ *Journal, Institute of Metals*, 1913, v. 9, p. 158.

structure was obtained, the metal as cast might fail under water pressure, owing to the fluid sweating "through the microscopical pores formed between the two constituents. These constituents are of widely different chemical composition, and possess different coefficients of contraction."

34. Primrose, in his 1913 paper, described a method whereby the porosity of gunmetal might be removed, namely, by annealing the material for 30 minutes at 1292 degs. Fahr. (700 degs. Cent.); but this treatment has been stated by some authorities to be dependent for its success, not on the structural change which Primrose found was produced in the alloy by heat treatment, but to superficial oxidation of the casting, which has the effect of preventing the seepage of water through the surface of the metal when under test. If the oxidized surface is machined away, the original troubles due to porosity reappear. It may be remarked that heat treatment at 1112 to 1292 degs. Fahr. (600 to 700 degs. Cent.), although it may not eliminate porosity, has the undoubted effect of improving the mechanical properties of the alloy.²⁸

35. Returning to the discussion of microstructure, Bailey found that "bars (both sand and chill) giving the best tests showed structures consisting of small isolated areas of eutectoid (α, δ and Cu_3P) in a ground mass of α ." This description of the ideal structure for bronze of this type is really the converse of that given by Primrose.²⁹

36. The ideal structure is obtained by the use of a high pouring temperature, the areas of eutectoid becoming larger and more connected as the pouring temperature is depressed. The coalescence of the eutectoid, which results from the use of low pouring temperatures, has a deleterious effect on the mechanical properties, as might of course be expected. However, Bailey found that by annealing for one hour at 1112 degs. Fahr. (600 degs. Cent.) bars which as cast contained their eutectoid in a continuous network, solution of much eutectoid resulted and the mechanical properties of the material were found to conform with those obtained in bars cast at more suitable temperatures. As Bailey points out: "This is not only of considerable practicable value in itself but confirms the . . . suggestions that faulty distribution of eutectoid is the sole cause of the weakness of material cast at low temperatures."

²⁸ Compare Bailey, *Journal, Institute of Metals*, 1923, v. 30, p. 401; see also paragraph 36 herein.

²⁹ *Journal, Institute of Metals*, 1913, v. 9, p. 158.

IX—MECHANISM OF INVERSE SEGREGATION AND OF THE FORMATION OF SHRINKAGE CAVITIES.

37. One is naturally led to inquire how the various phenomena described above can be rationally explained in terms of the behavior of a cooling casting. To start with, it may be well to emphasize the importance in this connection of a temperature gradient within a cooling casting. If the alloy, subsequent to its being poured into an open mold, were to remain uniform in temperature throughout at all stages during cooling, crystallization would start promiscuously throughout the entire melt and freezing would proceed in such a way as to produce a solid mass of the geometrical form of the mold, having a flat surface and a volume less than that of the molten alloy by an amount approximately equivalent to the volume contraction of the alloy on cooling from the pouring temperature to room temperature.

38. The effect of the mold, however, is to extract heat from the alloy at the mold-melt interface and to set up a temperature gradient within the melt. If the rate of transfer of heat across the mold-melt interface is low, the temperature gradient produced in the melt will be small. Given these conditions, crystallization will start at the mold-melt interface and will proceed inward, the solidifying alloy forming a container growing in thickness at the surface where solid and liquid make contact. Since the level of the liquid in this container falls slowly, due to the volume contraction of the alloy on freezing, the casting produced will be characterized by a conical pipe. Such portions of the alloy as are last to freeze will be concentrated in the body of the casting just below the pipe.

39. As the rate of heat transfer across the mold-melt interface rises, the temperature gradient within the melt and the tendency for peripheral dendrites to grow normally to the surface of the mold increase. The peripheral dendrites may continue to grow until they meet at the center of the mold. On the other hand spontaneous crystallization may occur throughout the melt in the central parts of the mold before the peripheral dendrites meet, in which case the shell formed by the latter may be looked upon as a mold in which the melt is freezing under conditions conducive to the formation of a pipe.

40. When the rate of transfer of heat across the mold-melt interface is sufficiently high, a skull may form over the top of the

casting, which may or may not effectively seal the interior of the casting from the air.

41. The conditions contributing to the production of high temperature gradients within the melt in a mold are such as lead to under-cooling and to the formation of small dendrites, both at the periphery and at the center of the melt. Given these conditions, the alloy in the mold, during the last stages of freezing may be imagined as filled with dendrites of which the limbs, branches, twigs and stems have already begun to interfere with one another, so that a system of capillaries, extending from the center to the skin of the casting and filled with tin-rich liquid, is formed. At any given time during the cooling of the casting the peripheral capillaries will be smaller than those near the center—on the one hand, because under-cooling is greater near the surface than at the center; on the other hand, because, generally speaking, the dendrites near the surface will be more developed than those near the center. Hence, just as the arteries and veins in the human body become smaller and smaller, the greater their distance from the heart, so the capillaries in freezing castings of alloys of this type become smaller and smaller, the greater their distance from the hottest parts of the castings.

42. The capillaries in such a system as this must not be imagined as tubes decreasing uniformly in diameter from the center to the outside of the casting, but as irregularly shaped, interconnected passages which lessen in peripheral dimensions, the nearer they are to the surface.

43. The peripheral dimensions of these capillaries might be expected to exert a considerable influence upon the movement of liquid through the system. Viscosity does, of course, play a part in retarding the movement of the tin-rich liquid through the system, but the fact that inverse segregation is the more pronounced, the greater the temperature gradient (*i. e.*, the finer the capillaries), seems to discountenance the idea that the viscosity of the tin-rich liquid is of real importance in this connection.

44. The tin-rich melt which fills this capillary system will tend to solidify first at the ends of the capillaries. However, the solidification of the melt at the ends of the capillaries would result in the formation of numerous small contraction cavities were it not that the production of these would involve the expenditure of an amount of surface energy considerably greater than that which

would be involved in the production of a single large cavity of similar volume elsewhere in the capillary system.

45. Since the surface energy per unit volume involved in the production of a single large cavity in the larger capillaries is less than that involved in the production of a number of cavities in the peripheral capillaries, the tin-rich liquid will solidify in the latter without the formation of contraction cavities, liquid meanwhile moving forward through the capillaries to make up for the change in volume resulting from the change in state.

46. Further, since the surface energy per unit volume involved in the formation of a cavity of given volume becomes less as the temperature rises, it is natural that such a cavity should form in the capillary system where the temperature is greatest, that is, at the center of the casting. The temperature gradient within the casting itself assists, then, in determining that the contraction cavities shall form in the body of the metal.

47. Another force tending to keep the peripheral capillaries filled with tin-rich liquid is that due to the difference in vapor pressure between the melt at the central and outer portions of the casting. Owing to its lower temperature, the vapor pressure of the tin-rich liquid in a cavity at the extremities of the capillaries would at all times be less than that of the liquid in a cavity in the body of the casting. This would be true whether the liquid in the body of the casting were open to the air (that is, were at atmospheric pressure), or whether the capillary system could be considered closed. Hence, the least tendency for the volume to change at the extremities of the capillaries (due, for example, to solidification of tin-rich liquid) would be immediately compensated for by the forward movement of the liquid in these capillaries under the vapor-pressure difference referred to above.

48. A capillary system filled with tin-rich liquid, such as has been described above, can be produced only when the temperature gradient within the cooling metal is very high, and when undercooling is pronounced (see discussion under Section II). The less the undercooling of the alloy, the larger will be the dendrites and the wider the capillaries; the less the temperature gradient in the cooling melt, the shorter will be the peripheral dendrites and the more uniform the peripheral dimensions of the capillaries throughout the casting. In a capillary system uniform in temperature and in peripheral dimensions, the chance that a contraction cavity will form in one part of the system—at, say, the extremities of the capillaries—will be just as great as the

chance that it will form in another, since the energy involved in the formation of either will be nearly if not quite the same, and the difference in the vapor pressure of the melt at different parts of the casting will be almost negligible.

49. It is suggested, therefore, that the phenomenon of inverse segregation is occasioned by the tendency for such contraction cavities as form in a rapidly cooling casting, in which a high temperature gradient exists, to do so in those portions of the capillary system where the least surface energy is required for their formation. The fact that the vapor pressure of the melt at the ends of the capillaries is always less than that at the center of the system, decreases the likelihood of the formation of contraction cavities in its more constricted portions. The result will be that the tin-rich liquid in the system will become concentrated in the peripheral capillaries, while the central capillaries become filled with contraction cavities.

X.—EFFECT OF METALLIC AND NON-METALLIC ELEMENTS AND COMPOUNDS ON THE MECHANISM OF INVERSE SEGREGATION

50. Alterations in the composition of the tin-rich liquid due to changes in the composition of the alloy will undoubtedly modify the behavior of the tin-rich liquid during the last stages of solidification. Hence, an alloy which, when cast under closely specified conditions, gives evidence of inverse segregation subsequent to solidification, may, when but slightly altered in composition by the addition of elements such as phosphorus, manganese or zinc, become quite normal in its behavior under the same conditions of casting.

51. The change in habit brought about by change in composition may be due to a number of causes, as, for example:

(a) The added element may change the temperature of primary solidification of the alloy, thus modifying the size of the dendrites and altering the ratio of their surface to their volume.

(b) The added element may alter the diffusivity of the alloy, thus modifying the temperature gradient of the melt within the mold.

(c) The added element may raise or lower the surface tension, or temperature coefficient of surface tension, of the tin-rich liquid.

(d) The added element may change the vapor pressure, or temperature coefficient of vapor pressure, of the tin-rich liquid.

52. The presence of solid, non-metallic impurities (for example, oxides and sulphides of the metals) in a casting is bound to affect the behavior of the tin-rich liquid during the last stages of solidification. Such impurities will tend to become concentrated in the capillaries and thus to interfere with the free flow of liquid through them. Anything tending to prevent the free flow of the melt in the body of the casting will, of course, lessen the tendency for inverse segregation to occur. Unless an ample supply of liquid is available during the process of solidification, which commences at the extremities of the capillaries, inverse segregation will be prevented.

53. As to the effects of dissolved gases on the behavior of alloys of the 89-11 type, it may be well first to refer to the work of Gayler³⁰ on the relationship between macrostructure, microstructure and pouring temperature in some nonferrous alloys. In this paper Gayler confirmed what is known to be true of bronze, namely, that the higher the temperature from which the alloy is cast, the coarser becomes the macrostructure and, to a lesser degree, the finer the microstructure.³¹ In the second place, she showed that, provided the casting temperature is low, the macrostructure and microstructure of an alloy do not seem to be affected by various gases. However, when the casting temperature is raised, the atmosphere to which the molten metal is exposed has a very marked effect on the macrostructure, together with a small effect on the microstructure. Hydrogen in particular, she found, caused the formation of a fine macrostructure in contrast with that obtained on casting under normal conditions.

54. Gayler's experiments were conducted on a copper-aluminum alloy containing 7 per cent copper. They proved, among other things, that after this alloy had been freed from gas by the nitrogen process and then melted in vacuum, it still showed inverse segregation. In reply to the discussion of her paper, Gayler sums up the results of these experiments in the following sentence: "Rapid cooling appears, therefore, to increase inverse segregation in the absence of gas; but when gases are present, the resulting effect of rapid cooling may or may not cause an increase in the segregation of the copper content."

55. It appears from Gayler's work that, when the casting

³⁰ *Journal, Institute of Metals*, 1930, v. 44, p. 97.

³¹ Papers by Rowe, *Journal, Institute of Metals*, 1924, v. 32, p. 73; and by Howard, *Foundry Trade Journal*, 1933, v. 48, p. 317.

temperature is raised, the gas to which the molten metal is exposed has a very marked effect on the macrostructure, together with a smaller effect on the microstructure. As suggested above, anything which modifies the size of the dendrites or alters their surface-volume ratio is bound to have an effect on the behavior of the tin-rich liquid in the capillary system which is produced in the casting during the last stages of solidification. If, for example, hydrogen has the effect of reducing the size of the capillaries in the system, the phenomenon of inverse segregation will be more marked than it would have been had hydrogen been absent.

XI—COMMENTS OF THE EFFECTS OF OCCLUDED GASES ON INVERSE SEGREGATION

56. Gases, although they may have no effect on the macrostructure and microstructure of the alloys in which they are occluded, may have an important influence on the behavior of the tin-rich liquid which fills the capillary system during the last stages of freezing.

57. The author finds it difficult to accept Genders' theory of the role of gases in producing inverse segregation, believing as he does that they play a secondary part only. It may be well to note that the natural place for gases first to evolve on cooling is in those parts of a casting which are first to freeze. If it be admitted that the capillary system visualized by the author exists in bronze and other alloy castings during the last stages of freezing, and that, in most cases, solidification of the tin-rich liquid occurs first in the peripheral capillaries of the system, why do not the gases concentrated in the tin-rich liquid come out of solution in the peripheral capillaries and by their expansion force the still-liquid tin-rich material toward the center of the casting? Such a process would have the effect of distributing the tin-rich liquid more or less uniformly throughout the capillary system and of preventing liquation. It may be that this is the course of events under certain conditions of cooling, but such a course of events would not produce the effects that we are trying to explain.

58. It seems more probable that liquation occurs when the gases occluded in the alloy become concentrated in the tin-rich liquid and do not come out of solution until the central parts of the casting are in process of solidification—with formation of contraction cavities. Now, if the difference in solubility of oc-

cluded gases in the liquid and solid tin-rich material is sufficiently great, these gases on coming out of solution may have an important influence on the behavior of such tin-rich liquid as occupies the capillaries of which the extremities are still unsealed by solidified material. If, for example, those portions of the capillary system which extend into the gates or risers of the casting are still filled to their extremities with tin-rich liquid, this liquid will be forced through these capillaries with the formation of the characteristic "button" of high tin content, of which a typical example has been shown in Fig. 2. In some cases the violence of the ejection of liquid from these capillaries may result in a veritable shower of metal, such as is sometimes seen above those parts of ingots and castings of certain metals as are open to the air.

59. The rates of cooling and degrees of under-cooling in different parts of a casting may in some instances be such that the solidification of tin-rich liquid, the formation of contraction cavities, and the evolution of occluded gas may start almost concurrently at the center of the casting, while the outer portions of the capillary system are still completely filled with tin-rich liquid. This case differs from that discussed in paragraphs 44 to 46, where solidification was presumed to have started at the ends of the capillaries, so that the capillary system was closed, that is, sealed from the atmosphere. In this case, the gas evolved at the center of the casting might exert pressure on the tin-rich liquid occupying the capillary system and, by pressing it out of the system, lead to the production of a skin on the casting of exceptionally high tin content, as is occasionally observed in practice. This is possibly the mechanism which leads to the formation of a skin of lead in bronzes containing this element.

60. The essential point to be borne in mind, however, is that in all instances where gases play a part in the movement of tin-rich liquid through the capillary system, their action is subsidiary to those of surface energy and vapor-pressure difference. The evolution of gases, the author believes, is always preceded by the solidification of tin-rich liquid in some portion of the capillary system, and such gas as is evolved can be effective in moving tin-rich liquid through the system only if the extremities of the capillaries at any part of the casting are still unsealed by solid material.

XII—CONCLUSION

61. The author hopes that the suggestions put forward in this paper may be of some value to those taking part in the discussion, and that the theory, when applied to practice, may lead to its improvement. Only by applying a theory to practice can its value be demonstrated. In conclusion, the author desires to express his appreciation of the assistance given him by Dr. J. N. Goodier of the Ontario Research Foundation, whose helpful discussion and criticism of the various points raised and suggested by the author have been of the utmost value in preparation of this paper.

(For discussion of this paper see page 377)

Symposium on Deoxidation and Degasification of Bronze Foundry Alloys

The Symposium on Deoxidation and Degasification of Bronze Foundry Alloys was held at the Convention of the American Foundrymen's Association on June 22. T. E. Kihlgren, International Nickel Co., Bayonne, N. J., acted as Chairman and Dr. C. H. Lorig, Battelle Memorial Institute, Columbus, O., as Vice Chairman. The meeting was conducted as an informal discussion group. The subject was divided into the three following types of alloys:

Alloy Group 1: Valve Bronzes, 85-5-5-5 type; discussion led by John W. Bolton, The Lukenheimer Co., Cincinnati.

Alloy Group 2: Tin Bronzes, 88-11 type; discussion led by O. W. Ellis, Ontario Research Foundation, Toronto, Canada. Mr. Ellis read a paper on "Mechanism of Inverse Segregation," presented on pp. 347 to 369 inclusive.

Alloy Group 3: Leaded Bearing Bronzes, 80-10-10 type; discussion led by G. H. Clamer, Ajax Metal Co., Philadelphia.

VALVE BRONZES—85-5-5-5 TYPE

J. W. BOLTON: The term deoxidation and degasification is a broad one. In the past, the term deoxidation was used to cover the causes of difficulties which result in leaky castings. After testing, such castings usually show discolored fractures. The microscope indicates that there is a discontinuity between the grain boundaries that explains the inability of the castings to withstand pressure.

For many years, deoxidizers have been used to overcome these and kindred troubles. If we go back a few years, all troubles in bronze castings apparently were laid to improper deoxidation. Remedies and nostrums advocated were without number.

Recently, I have re-read H. C. Dews' book, "Metallurgy of Bronze," in which he describes in detail the founding and crystallization characteristics of Admiralty metal, the familiar 88-10-2 zinc alloy. The British lay tremendous stress on pouring tem-

NOTE: This symposium was held as a part of the Nonferrous Division program at the 1933 Convention of the American Foundrymen's Association.

peratures. If you will read the book and analyze it carefully, you will find that over a certain narrow range of pouring temperature, under any reasonable conditions of foundry melting practice, a fairly sound alloy can be produced. The object is to get the temperature just high enough to pour the metal and yet low enough so that when poured into the mold the freezing rate will be so rapid that the possibility of gas rejection is prevented by quick cooling through a critical range.

Formerly, most nonferrous foundrymen felt that the maintenance of a reducing atmosphere was the best furnace practice. Today, they are emphasizing neutral atmospheres without defining them. One definition is that a neutral atmosphere does not contain excess carbon monoxide, hydrogen, hydrocarbons, oxygen or oxidizing gases. A few years ago in the New York meeting of the Institute of Metals someone suggested that a neutral atmosphere is one that is without effect on the properties of the metal. That possibly is the best definition.

QUESTION*: WHAT IS YOUR OPINION OF THE MOST SATISFACTORY METHOD OF DEOXIDIZING THE 85-5-5-5 ALLOY?

We like to melt under a neutral condition from a combustion gas point of view. We desire to reduce the carbon monoxide, hydrogen, hydrocarbons, and oxygen in the products of combustion to a minimum, regardless of whether the melting is done in a gas or electric furnace. We usually add a small percentage of phosphorus. We find that final phosphorus content should be only a few thousandths of one per cent.

QUESTION: WHAT OTHER DEOXIDIZERS HAVE YOU FOUND USEFUL?

Silicon is a fine deoxidizer in certain types of alloys but should not be used excessively in the 85-5-5-5 alloy. A few hundredths per cent can be used without much trouble. There are other deoxidizers, such as calcium and boron, which might be considered. Phosphorus remains the most common and safest deoxidizer.

QUESTION: WHAT IS THE EFFECT OF SUCH METAL ADDITIONS AS ZINC, PHOSPHORUS, MANGANESE, SILICON, BORON, MAGNESIUM, CALCIUM, BARIUM, ALUMINUM, ETC? HOW DO THEY COMPARE IN USEFULNESS? WHAT ARE THEIR LIMITATIONS, ETC.?

It is recognized that a small percentage of zinc is beneficial in most of these alloys. Consider Admiralty metal. It is much harder

*Questionnaires containing the questions in this discussion were mailed to members of the Nonferrous Division of the A.F.A. prior to the meeting.

to make an alloy of 80 per cent copper and 12 per cent tin without any zinc and phosphorus than it is when a small percentage of zinc is used in the alloy.

In the case of nickel alloys, such as monel metal and others of that type, magnesium has proved useful. We do not know the results obtained when it is used in red brasses.

Calcium should hold some promise.

Some salts of barium have been used, such as sulphates, to reoxidize an over-reduced metal to prevent difficulties that may arise from an over-reducing atmosphere.

Aluminum, which finds favor in steel metallurgy for preventing oxidation in cast metal, proves to be a boomerang when used in appreciable amounts in bronzes.

QUESTION: WHAT CASTING DEFECTS ARE MOST COMMON IN THE TYPE OF ALLOY UNDER DISCUSSION?

Some of the worst foundry difficulties are the result of metal that has become gassed due to too high a percentage of various reducing gases in the products of combustion. Incidentally, when bronzes are melted under conditions that are not extremely reducing, I have observed that contraction shrinkage is considerably greater and heavier gates and risers are required.

QUESTION: WHAT FURNACE ATMOSPHERES ARE MOST, AND LEAST, DESIRABLE?

When metal is melted in a neutral atmosphere, that is an atmosphere containing no reducing gases or free oxygen, and a small percentage of phosphorus is used, the range of pouring temperature is much greater than it is when no atmospheric control is exercised. Much research has been done both in the United States and abroad on pouring temperatures. On examination, it will be found that pouring temperatures are regulated to get the material to as low a temperature as possible so that it will cool quickly to avoid rejection of gas, yet have it flow and feed properly. That is not a preventative of the basic cause, namely, gasification during melting.

WRITTEN DISCUSSION

P. BARKER:¹ The following statements with respect to a "Symposium on Deoxidation and Degasification of Bronze Foundry Alloys," are brief and somewhat general. The subjects are so comprehensive that it would take many pages to cover them properly.

The most suitable deoxidizer is phosphorus; for 85-5-5 and 80-10-10,

¹ Federated Metals Corp., Detroit.

15 per cent phosphor-copper and for tin bronze of the 89-11 type, phosphor-tin. These deoxidizers have the advantage of performing their function without leaving a residual hardening and embrittling agent, when applied in the proper proportions.

Other deoxidizers used are silicon, manganese, calcium, magnesium, boric acid, boronic alloys and sometimes zinc and barium, etc. Special trade hardeners are on the market, containing deoxidizing agents for example, such as calcium. Some foundries find these particularly adaptable to their particular conditions. We have used successfully 15 per cent phosphor copper, 25 per cent manganese copper, 10 per cent silicon copper and boric acid for deoxidizing purposes.

Deoxidizers are essential to overcome the oxidation conditions of casting. Foundry conditions such as alloy used, type of melting furnace, nature of casting defect to be overcome, conditions of casting, such as distance of travel from furnace, length of time before pouring, kind and size of casting, etc., will determine the kind and amount of deoxidizer to use. In some instances, it may be dispensed with, while in general, the intelligent use of a deoxidizer is both beneficial and necessary as it is impossible to melt down in practical commercial foundry applications without some oxidation.

In general, deoxidizers act as hardening agents. They increase the tensile strength and decrease the percentage of elongation. This does not necessarily follow with each deoxidizer under consideration. Some act in greater or less degree than others. The object is to rid the casting alloy of oxides which accrue during the melting down process and which cause unsatisfactory castings. The use of excessive amounts of a deoxidizer resulting in a residuum, which frequently causes brittleness and cracks. Aluminum causes leaky castings as it forms weak areas at the intergranular boundaries.

The most common deoxidizers are phosphorus, manganese, silicon, zinc, calcium, magnesium. The usefulness of boronic compounds as deoxidizers has been disputed but the writer has used boric acid successfully as a deoxidizer for copper over a period of months.

In general, a deoxidizer which performs its functions of ridding the alloy of excess oxygen and oxides and then volatilizes or rises to the surface as a slag, whence it may be readily removed, has a preference over one which remains as a compound in the alloy. Phosphorus, manganese, silicon, magnesium, and zinc generally are more desirable than calcium, barium, aluminum. Care must be exercised to keep the excess down to a minimum. Use of too much deoxidizer causes brittle castings. Oxides and slag must be kept to a minimum for sound and clean castings.

The use of properly controlled remelt ingot is a decided factor in obtaining satisfactory castings as the successive remelting causes closer alloying of the separate ingredients. Sulphur in conjunction with other impurities in commercial alloys is objectionable, as are all elements which gasify and cause porosity. Excessive iron causes hard spots and poor machining qualities.

Casting defects most common in the previously mentioned alloys are porosity and cracking. In melting down and casting, care should be exercised to exclude oxidation, or keep it at a minimum. Impurities

which gasify, such as sulphur, should be kept low or entirely eliminated.

To obtain gas-free metal, melt under cover of borax or glass, etc. Pour as soon as possible after skimming. Pour as near gate as possible, and in some instances have a neutral or reducing gas in the flask. Do not overheat and burn the metal.

A neutral atmosphere in the furnace is the most desirable and an oxidizing atmosphere the least desirable. Since an absolutely neutral atmosphere is difficult to obtain in most furnaces a very slightly oxidizing atmosphere generally is used for economical fuel operation and the metal is kept covered.

ORAL DISCUSSION

R. L. BINNEY:² When I think of a gassed bronze, I think of a bronze containing occluded gases, gases trapped physically. Likewise, an oxidized bronze is one in which an oxidizing reaction has taken place that leaves the metal charged with oxides.

H. M. ST. JOHN:³ My impression of an oxidized metal is one which is sluggish due to dissolved impurities. In some cases, a portion of these impurities may really be deoxidized but if the metal can be made more fluid by the addition of phosphorus, the final product may be classified as deoxidized metal. Although I am not sure that the metal in question actually is in a deoxidized condition before the phosphorus treatment, I am convinced that it had not been deoxidized by the furnace atmosphere. In this sense, I feel sure that carbon monoxide is neutral in brass-furnace atmospheres.

To satisfy myself thoroughly on this point, I have used a small laboratory electric furnace which, if allowed to operate in its own way, would contain an atmosphere of about 20 per cent carbon monoxide. I have injected carbon monoxide into the furnace atmosphere until tests showed it to be approximately 99 per cent carbon monoxide. The metal poured from the furnace, when the highest percentage of carbon monoxide possible in the atmosphere was used, was as sound as that poured from a 20 per cent carbon monoxide atmosphere or as metal poured from furnaces where the carbon monoxide in the atmosphere has been reduced to as low a point as possible.

I am inclined to believe that carbon monoxide is neither injurious nor beneficial except for the fact that when you have carbon monoxide present you do not have oxygen. If oxygen is needed as a refining agent, the carbon monoxide is objectionable because it takes the place of the desired oxygen.

In the following reference to discontinuity of structure, I am not referring to what is called sponginess, that is, spherical openings in the metal which are bubbles, but to an angular discontinuity, crystals which are separated or forced apart and have fissures between them. The latter discontinuity of structure occurs sometimes in castings and must be due to some cause other than the presence of gas. As far as I have been able to discover, there is no connection between this discontinuous condition and the type of atmosphere in the furnace.

² Binney Castings Co., Toledo, O.

³ Detroit Lubricator Co., Detroit.

One point in connection with gassed metal that deserves more consideration than it has had in the past, is the condition where sponginess actually does exist. The small, round holes usually are as bright as new minted gold, indicating the presence of a reducing gas. The possibility that this reducing gas is hydrogen, is something which needs further investigation. It seems to me that hydrogen is the cause of that type of gas defect.

J. W. BOLTON: Facts in our researches showed that under reducing conditions, as previously defined, porosity resulted and under neutral gas conditions it did not. The majority of investigators check our results and several go as far as to state that slightly oxidizing atmospheres are desirable.

In porous metal, fissures actually are found between the crystals. They formerly were termed oxides. They are nothing but holes. The term "incipient shrinkage" was coined because the same phenomena could be found in castings insufficiently fed. Incipient shrinkage means the beginning or start of shrinkage.

Dissolved gases, carbon monoxide, hydrogen and like other gases, are evacuated during solidification and are found in the crystal interstices. Neither Mr. St. John nor I actually have identified the gases nor has anybody else proved the case. Some extremely tedious and painstaking work has been done to determine what gases are in gassed metal but there has been no satisfactory, full explanation of the connection between the gases found by analysis and those to which troubles might be attributable. Contraction and lessened gas solubilities are the likely sources of trouble. In some cases, shrinkage alone may be responsible, in others gases are the more important factor.

H. W. MAACK:⁴ Referring to the use of phosphor-copper as a de-oxidizing agent, with improper metal condition, phosphor-copper will cause segregation. If the metal condition is as it should be, an amount of phosphorus in excess of that necessary to oxidize the bronze will not have such harmful effects. Pouring temperature suitable to the cross-section exerts a great influence on the quality of the casting as judged by fracture. However, I think that phosphor-copper often is inclined to cause segregation.

The requirements of a proper melting condition are not clear to me. It is my opinion that bronze melted in a pitch-coke-fired cupola is of extremely high quality. With that type melting unit, there probably is a sequence of conditions strongly oxidizing in the upper zones of the cupola and reducing as the metal passes through the incandescent coke. Possibly the oxidation and reduction is responsible for the high quality of the resultant metal. Whether it is due to the elimination of traces of silicon and aluminum, which would be harmful, or to other effects, I do not know.

J. W. BOLTON: There is one analogy that often has been used by opponents of those who hold the same views as those of Mr. St. John. That is the question of the difference between over-poled copper, over-reduced copper and tough pitch copper that contains a certain percentage

⁴ Crane Company, Chicago.

of oxide. The difference between over-poled copper and those that contain a certain amount of oxide is very marked.

SAM TOUR:⁵ I cannot conceive of free oxygen being present in a bronze. Copper forms copper oxide rapidly and copper oxide is considerably soluble in molten copper. The oxidized molten metal would be expected to be a solution of copper oxide in copper. It is difficult to imagine that any oxide in addition to copper oxide would be present in any great quantities. Silicon would be oxidized but silica would not be expected to dissolve in the copper. Tin would be oxidized to tin oxide but it would not dissolve. The same possibly is true of lead.

I picture an oxidized bronze as one which contains oxygen in solution as copper oxide. That would be a modification of the alloy. An oxidized bronze might be one which contains particles of oxidized metals of other kinds such as tin oxide, silica, etc. In the latter case it would be called a dirty metal since it is a metal containing non-metallic inclusions in the form of oxides other than copper. If the metal is oxidized to the state of dirtiness, some agent, such as phosphorus, should be added to make the metal more fluid, and allow the oxides to float out of the bath. If the metal contains copper oxide in solution, it should no longer be considered an 85-5-5-5 alloy. Another constituent, copper oxide, is present and has modified the alloy.

If reduction of the copper oxide in solution is attempted by the method used in refining copper by poling, that is by the introduction of hydrocarbons, which will reduce the amount of oxygen in the solution, it can be expected normally that the oxygen will be reduced by the formation of carbon monoxide at the high temperatures of molten copper. Not much carbon dioxide would be formed. The carbon monoxide will pass off as a gas.

Indications from that thought would tend to confirm Mr. St. John's conclusion that carbon monoxide is not detrimental to copper alloys. Carbon monoxide alone is not a bad actor. Carbon monoxide present in the atmosphere can affect conditions depending on its reaction with other gases that may be present in the atmosphere. If hydrocarbons are present with carbon monoxide, the hydrocarbons can enter the metal because there is no free oxygen present to react with them. Instead of oxygen, I think the difficulty is largely dissolved gases other than oxygen and carbon monoxide. The difficulties probably are due to the presence of hydrogen, methane and possibly some metal hydrides.

Silicon is known to form a hydride and probably other metals also do. Thus, gassed bronzes contain hydrogen either as such or as a compound with the metal, the solubility being much greater in the molten than in the solid metal. As the molten metal begins to freeze, gases are rejected and result in swelled gates, spewing out of the gates, etc. Certainly a spewing gate cannot be called oxidized metal, yet it often is treated as such and phosphorus is used to eliminate the gases. When metal is treated to eliminate oxygen, some element or compound is used that will react with the oxygen. When metal is treated to eliminate gases, possibly something should be added that will react with the gases.

It is interesting to find that gases often can be eliminated by another

⁵ Lucius Pitkin, Inc., New York.

gas. In the aluminum field, for example, nitrogen bubbled through molten aluminum will remove considerable quantities of dissolved gases. It is entirely possible that the addition of phosphorus will degas bronze, not by reaction with the gases but by disturbing the equilibrium between the gas and the base metal thus causing the gas to be released and to rise to the surface.

TIN BRONZES—88-11 TYPE

O. W. ELLIS: When I was invited to lead the discussion on these alloys, Mr. Kihlgren suggested that I prepare a paper. My paper deals more particularly with the physical and mechanical properties of alloys of the 88-11 type because I felt that we could not discuss logically the effects of gases and oxides on the alloys if we did not have some clear ideas about the alloys themselves. (Mr. Ellis' paper will be found on pages 347 to 369 inclusive.)

ORAL DISCUSSION

G. H. CLAMER⁷: The reference that you made to my observations of some years ago to what we call eutectic sweat in our foundry, is direct evidence of the capillary action you have described. We had been in business for many years and occasionally received complaints that castings were unsatisfactory because they had a surface hard to machine. An investigation showed us that the hard skin had a much higher tin content than the body of the casting. The problem was, how did that hard skin containing a higher tin content than the casting get there? According to the copper-tin diagram, the high tin constituent should be in the center of the casting, the part that solidified last. Such was not the case. The hard skin contained from 17 to 22 per cent tin and the body of the casting only 10 per cent.

In casting ingots of 80-10-10 metal, particularly those containing phosphorus, we frequently find that after the ingot has cooled below a red heat, this high tin alloy oozes out the capillaries and shows itself on the surface of the cold ingot. If those nodules are broken off, it will be found that they are of a higher tin content than the ingot.

I once examined a locomotive driving brass that had been heated in service. The usual oil hole was at the top. Under the effect of heat and pressure, the high tin constituent had squeezed out and had run into the oil hole. In that case, the tin content of the nodule taken from the oil hole was as high as 22 per cent. The particular bearing was phosphor bronze and normally should have contained about 0.7 per cent phosphorus. This nodule was found to contain 3.0 per cent. This illustrates the tendency toward the formation of the eutectic alloy and also the tendency of that alloy to segregate, inversely under certain conditions of temperature and pressure.

⁷ Ajax Metal Co., Philadelphia.

We discovered that it was necessary to limit the impurity content in such alloys. Segregation of the high tin constituent was greater as the proportion of zinc and antimony was increased.

E. F. HESS⁸: We make a casting from an alloy that falls in this group, although it is not exactly the same. We add a little lead for machining purposes. At different times we have run into this form of segregation, dendrites or whatever it may be, in different parts of the castings. I ran a series of tests varying the phosphorus content thinking that I could get rid of some of the impurities that looked like oxides. They were immediately under the skin and would show up after the castings were machined.

With a very small amount of phosphorus, the castings looked nice. By increasing the phosphorus content up to a point where the castings became a white color, we found that the tin content ran about 17 or 18 per cent. The skin was extremely hard and a protest was made about the difficulty of machining. As a result, I have tried to keep the phosphorus content low, although the high phosphorus metal is nice under the skin.

The casting in which the difficulty developed was roughly a solid stem about 1 inch in diameter and 4 inches long joined to a heavy ring about 4 inches in diameter and 2 inches thick with a cored out place in the stem side. Various amounts of phosphorus seemed to influence the porosity of this particular casting quite differently.

When a small amount of phosphorus was used, there was porosity through the heavy ring at the stem. This was made visible by pouring kerosene into the depression and watching it come through after several minutes. The porosity was overcome by raising the phosphorus content.

O. W. ELLIS: What I tried to point out in my paper was that porosity can be explained in terms of physical phenomena only. It is necessary first to imagine a system of capillaries and next to ask why it is that in such a system there is a tendency under certain conditions for the central portions to be porous and the outside to become rich in tin. The reason, I opined, must lie in the fact that it is easier to produce blow holes or cavities in the center of the system than at the outside. Why is it easier to produce cavities at the center than at the outside? Because the amount of energy required to produce one large cavity at the center of the system of capillaries is less than that needed to produce, say, 20 small cavities at the ends of the capillaries.

Contraction cavities which frequently are referred to as blow-holes and which, without doubt, contain gas, are present, not because of any chemical phenomenon but because of a physical phenomenon. They contain gas, but the gas does not enter them until they are in process of formation, that is, until the last traces of tin-rich liquid begin to throw out the gases they hold in solution. It may be that when a large amount of gas comes out of solution in the tin-rich liquid, it will push its way against the relatively mushy material at the top of the casting and produce the phenomenon of tin sweat.

SAM TOUR: If we picture a rectangular section crystallizing from

⁸Ohio Injector Co., Wadsworth, O.

the outside in, with dendrites all in tapered form, assuming they are symmetrical, leaving capillaries which are smaller at the surface than at the center between these dendrites, we have space left for the liquid increasing to the center whereas the space occupied by the dendrites increases to the edge.

If what Mr. Ellis proposes is true, those initial dendrites, as they solidify, will be lower in tin than the original melt. Then we would have the surplus at the surface lower in tin than the center liquid. Then due to capillary action, the high tin liquid would run back through the capillaries alongside the dendrites. Since the diameter of the capillary decreases as it nears the surface, there would be less of this tin-rich liquid to get to the surface than there would be somewhere in between so that you would not have tin enrichment on the surface of the casting due to the molten liquor running back into the dendrite unless something back of that were pressing it further.

If the amount of liquid left is not sufficient to fill the entire central cavity, which is what we expect due to shrinkage, then we would expect the central cavity to be empty and the capillaries full of tin-rich liquor.

If an analysis was made and Mr. Ellis' theory was correct, low tin would be found on the surface, high tin half way between the surface and the center and low tin in the center. The liquor has not been sufficient in volume to fill the center. It has been pushed to the outside and has stopped somewhere between the outside and the center. Thus we have a low tin surface, high tin next and low tin center.

That does not account for tin sweat. Since the capillaries are smaller at the outside near the surface, tin sweat cannot be accounted for unless something forces the tin-rich liquid out through the capillaries onto the surface. That is where the rejected gases enter the problem. Gaseous metal will sweat and such sweat may, therefore, indicate gassed metal. If the center of a casting could be held open and exposed to the air during the period of gas rejection and the gases liberated at the center of the casting were allowed to vent, tin sweat would not take place.

Gases dissolved in metal occupy little, if any, volume but when they come out of solution, they occupy enormous volumes. As the metal in the center is freezing, it is liberating gases and creating considerable pressure. That pressure can get to the outside only by first pushing the tin-rich alloy out ahead of it. The tin-rich alloy breaks through the end of one of the capillaries and forms a mushroom growth on the surface in the form of tin sweat.

W. E. PAULSON*: I think Mr. Tour overlooks the fact that a capillary is a small tube and the smaller the tube the greater the tendency for the liquid in it to flow as far as it can. I think that the fact that the tubes become smaller as they reach the surface would explain why the tin-rich liquid might come to the surface easily in a tube that became gradually smaller as it approached the surface of the casting. A large tube does not have capillary attraction to any great extent.

O. W. ELLIS: I have not suggested that capillarity comes into play at all. As a matter of fact, if two capillary tubes, one uniform in

* Thomas Paulson & Son, Inc., Brooklyn.

diameter and the other gradually decreasing in diameter, are compared, the heights to which liquid will rise in both will be exactly the same. Only the diameter of the tube determines the height to which the liquid will rise.

Mr. Tour has pictured a much coarser system than I had in mind. I had thought of a system of extremely fine capillaries. The tin-rich liquid at the last stages of freezing occupies not more than 6 per cent of the total volume of the alloy. Whenever inverse segregation occurs, porosity also is present. Therefore, it seems that the fact the central portions of castings are porous must be connected with the enrichment of the tin on the outside.

LEADED BEARING BRONZES—80-10-10 TYPE

G. H. CLAMER: What is true of the 85-5-5 alloy holds also for the 80-10-10 alloys. One vital consideration in the 80-10-10 group is that there are two distinct 80-10-10 alloys provided for in the specification of the A.S.T.M. These alloys act differently and are being handled separately in the recommended practice sheets of the A.F.A. I refer to the 80-10-10 alloy with zinc present and the same alloy deoxidized with phosphorus or with an excess of phosphorus present. They must be handled by different methods in the foundry. The 80-10-10 type referred to for discussion probably includes the general classification of bearing metals which may contain 15 per cent or more lead and 6 per cent or more tin.

Referring to Mr. St. John's information that carbon monoxide should not be classed as a reducing but as a neutral gas, if that statement is true, it would seem that the use of charcoal, which is used generally for preventing oxidation, might be dispensed with, except that it keeps the air away and in that way prevents the metal from becoming oxidized.

All metals listed as deoxidizers are electropositive to copper and therefore are more easily oxidized than copper. That does not mean necessarily that there may not be some copper oxide in such alloys. Reactions do not proceed instantaneously and they frequently are not entirely completed. I am not prepared to say whether copper oxide really is present if zinc or other highly electropositive metals are contained in the alloy.

H. C. Thews states that phosphorus will reduce zinc oxide. That would explain satisfactorily why the addition of only a small amount of phosphorus to the 85-5-5 alloy or the 80-10-10 alloys containing zinc is beneficial. We always have known that it made the metal more fluid without temperature elevation. If phosphorus does reduce zinc oxide, this satisfactorily explains its beneficial effect.

ORAL DISCUSSION

O. W. ELLIS: Three or four years ago, I read a paper before the American Institute of Mining and Metallurgical Engineers on the subject, "Oxides in Brass." We made a number of different melts in an oil burning furnace, using conditions which were as nearly identical as possible but varying the character of the charge. In some cases we used all metal charges; in others, scrap plus metal charges; in another ingot charges and in other cases the worst scrap that we could find in the foundry. We took samples from these charges and poured them into standard molds.

The first point noted was that the amount of oxide varied with the character of the charge. We determined the amount of oxide by measuring the size and counting the number of oxide particles microscopically.

We found that where the metal was held in the furnace under what would normally be termed reducing conditions for a period of about 2 hours, the amount of oxide present was practically the same after this long heating period as it was after a considerably shorter time. In other words, keeping the metal at rest under a reducing atmosphere for what might be described as an exorbitant length of time had no effect upon the oxide content of the alloy. Incidentally, the alloy was the type used in the manufacture of trolley ears.

An interesting point we discovered was that when the alloy was poured into sand molds, the particles of oxide in the castings were smaller than when the alloy was poured into chill molds. This indicated clearly that zinc oxide actually is soluble in the alloy in the molten state and that the size of the oxide particles is determined by the rate of cooling.

We found no relationship between the mechanical properties of the alloy and the amount of oxide present.

H. M. ST. JOHN: I liked Mr. Bolton's definition, that a neutral atmosphere is one which has no effect on the metal which is being handled. There is no question in my mind about the reducing character of the carbon monoxide when dealing with straight copper which contains copper oxide. It unquestionably is reducing.

If there is any copper oxide present in a melt of the type brass or bronze under discussion, it is possible that this also is reduced by the carbon monoxide. However, if so, it is unimportant because the amount of dissolved oxide present must be extremely small and its effect on the character of the metal practically nil. The important point is that the presence of solid carbon or hydro-carbon on a melt of valve bronze will reduce silicon from silica and seriously injure the metal. Possibly other reductions of a similar character take place.

In melting bronzes in a furnace which normally has an atmosphere high in carbon monoxide, the presence of solid carbon or oil in contact with the metal should be avoided. If this is possible and the metal is free from contamination at the start, the presence of carbon monoxide is quite harmless.

W. E. PAULSON: Do you consider carbon as contained in oil in the same classification as that present in charcoal?

H. M. ST. JOHN: You are thinking probably of melting in crucibles where certain amounts of oxygen are present. I am thinking more particularly of the type of electric furnace in which the atmosphere of the furnace contains no oxygen at all. When oxygen is present, oil is much less detrimental because the reduction does not take place in the presence of oxygen.

G. H. CLAMER: Charcoal is used much more than is necessary. Frequently foundrymen throw charcoal on babbitt metal that has not reached a red heat. Under such conditions, the charcoal is wasted for it cannot perform any helpful function.

I referred to the use of charcoal on the bottom of the ladle and the gas condition that resulted when the metal was poured over it. If gassed conditions under such circumstances do not result from carbon monoxide, what do they result from? The explanation may be water vapor originating from water in the charcoal. The only way to determine the nature of the gas is to use the method that has been devised by the Bureau of Standards, that is, melting the alloys under vacuum conditions and determining the actual composition of the gas.

E. F. HESS: Nothing has been said about the metals used. I refer especially to copper. We had some trouble in melting a certain batch of copper. It was melted as received, gating, etc. was exactly the same as in other instances, yet suddenly our percentage of leakers increased considerably. I noticed in this lot of copper a considerable amount of reduced copper ingots. Heats of those ingots melted in the carbon monoxide atmosphere proved to be the cause of the leakers because when we used flat top ingots, the number of leakers returned to normal.

In the 85-5-5-5 alloy we used large gates. The size of the gates was not increased and on leaky castings, there was no shrinkage from the gate to the casting. It seemed to come from the lighter and heavier sections of the casting, a sort of inward shrinkage.

E. F. CONE¹⁰: What is the recommended practice regarding the time of introduction of phosphorus in the open pit and crucible in the 85-5-5-5 alloy? How can you get it down to the bottom of the crucible? It seems to work to the surface.

G. H. CLAMER: Phosphor-copper usually is added just before pouring. Do not pour too quickly as time must be allowed for the reaction to take place. The ordinary way of introducing yellow phosphorus is by means of a phosphorizer, which is an inverted crucible attached to an iron bar, and plunging the phosphorus below the surface.

MEMBER: In the written discussion, the statement is made that deoxidizers reduce elongation since they act as hardeners and increase the tensile strength. That may reduce elongation but the justification for the use of deoxidizers is improvement of the metal. The statement also is made that continuous remelting improves the alloy. Personally, I doubt that statement. In a paper which I presented several years ago before the Engineering Society in Montreal, a great deal of work was done because of a question that arose frequently in the selling end of the foundry business. Our castings were made from new metal while our com-

¹⁰ New York.

petitor's were made from scrap. He boasted that because the metal had been melted so many times it was much better. We found, after making the alloy carefully with new metal, that we got as good results as with the first and second remelting of the virgin metal.

Concerning the use of deoxidizers, if two foundrymen use a deoxidizer under practically the same circumstances, one will say that it is good and other will hold the opposite opinion. In the first case, sufficient time was allowed for the deoxidizer to react and in the other it was not. That time requirement has not been given sufficient attention in the average foundry. A higher melting temperature must be used to allow for the required time. The method of stirring also exerts an influence on the alloy. I have found many cases where the deoxidizer is stirred into the metal and its purpose is defeated by stirring it into the dirt in the bottom. The idea is to mix the deoxidizer with the metal and avoid the dirt.

Referring to burned metal, foundrymen are told not to burn their metal. Is oxidizing what is meant by burning? If so, how can some castings be poured at 2000 degrees Fahr. and others at 2300 degrees Fahr. without burning the metal?

W. EMBER¹¹: I do not know whether what I would call burned metal would be called by the same name by foundrymen in general or not, but we have found a condition in our foundry which we call burned metal. We make a bronze containing 9 per cent tin, 4 per cent zinc and the remainder copper, which might be classed in the 88-10-2 type. We have found that we can overheat that metal, meaning that the temperature is in excess of 2400 degrees Fahr. and cool it to what we think is the correct pouring temperature. When that procedure is followed, the resulting casting as it comes from the mold has the appearance of a sheet of galvanized iron. In many instances it is possible to detect with the naked eye, surface cracks between each of the crystals with the galvanized sheet iron appearance. That is what we call burned metal. We can make that condition occur by overheating the metal but we cannot always prevent it from occurring.

When I say the casting has the appearance of galvanized sheet iron, I do not mean that it has a white color. It has the red color of the metal but the surface has the enlarged crystalline form found on galvanized sheet iron. The correct pouring temperature is 2150 degrees Fahr.

G. H. CLAMER: Mr. Ellis, do you think the appearance is caused by enlarged dendrites on the surface?

O. W. ELLIS: It might be that very high pouring temperatures change the form of the capillaries. However, it seems to me that some material has entered and has spoiled the alloy. When melting at a very high temperature, it might be possible for a reaction to take place between the pot and the metal with the result that a new constituent, not normally present, would enter the alloy.

MEMBER: At what temperature is the metal free from this defect?

W. EMBER: I do not know at what temperature it does not occur. Our pyrometer reads only to 2400 degrees Fahr. and I know the temperature of the metal is above that. We try to keep the metal not over

¹¹ Jefferson Brass Foundry, Brooklyn.

2400 degrees Fahr. When that temperature is not exceeded we do not have the trouble mentioned.

G. H. CLAMER: Did that trouble appear on the skin side and not on the machined surface?

W. EMBER: The trouble appeared throughout the metal as far as we could find. What I wanted to emphasize is that the cracks are visual. If the casting is broken, the cracks can be traced all through it. If the casting is machined, it is possible to see where the machine tool has turned up the edge of the crystals. It is possible to see a definite division between the crystals.

O. W. ELLIS: In connection with the remark I just made, in making melts of cast iron in magnesia crucibles, it is possible to reduce magnesium from the pot. Might it not be possible at very high temperatures for a reaction to take place between a melt of bronze and a silicious crucible or furnace lining with the formation of silicon? If the reduced silicon entered the melt, it would spoil the alloy.

G. H. CLAMER: The impurity content of the alloy may be an explanation.

W. EMBER: The alloy was made from L.N.S. brand copper, straits tin and Horse Head zinc. All were purchased in the original slabs and pigs.

Studies on Solidification and Contraction and Their Relation to the Formation of Hot Tears in Steel Castings *

BY CHARLES W. BRIGGS† AND ROY A. GEZELIUS,†
WASHINGTON, D. C.

Abstract

The following studies, resulting from a compilation of data and expressions, consider the relation of solidification and contraction to the formation of hot tears. Casting solidification involving the principle of equal cooling throughout is undesirable. Such conditions should be replaced by the more natural method of directional solidification. Controlled directional solidification is successful only when certain principles of casting design are adopted. These principles stress the importance of feeding smaller sections through heavier ones and by studying mass effect by the use of inscribed circles. Metal contraction and hindered contraction resulting from mold resistance is responsible in the major part for the existence of hot-tear cracks in steel castings. Other important factors are cavities and the physical properties of the steel. There are two types of hot-tear cracks, (a) the internal hot tear that develops during the solidification of the casting, and (b) the external hot tear that develops after solidification is complete, but while the temperature is about 1300 degs. Cent. Intelligent application of certain principles in casting design is one of the most important of seven methods that may be adopted to obviate the danger of hot tears, or at least limit the hot tear formation.

1. There are many conflicting opinions on all phases of casting procedure and various remedies for all the numerous problems arising in the industry. These remedies often have been

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† Division of Physical Metallurgy, U. S. Naval Research Laboratory, Bellevue, Anacostia, D. C.

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applied by rule of thumb or as experience has dictated their use. It is surprising to note how little experimental data are available on the actual physical processes which occur during the solidification and contraction of steel castings.

2. It is not possible to review as large a field as the entire casting process in one article and do justice to all of its phases. The authors have selected, therefore, only one of the important phases, that of solidification, contraction and the formation of hot tears, and have attempted to bring together the best of available data and to supplement them with the results of some experiments.

3. In the industry today there are so many definitions for the several terms that in order to clarify the discussion it has been found necessary to define certain of these terms.

DEFINITIONS

Contraction, Shrinkage.

4. This is a phenomenon, occurring in most metals, whereby a decrease in temperature is accompanied by a decrease in volume. The change may be divided into three types, (1) the decrease in volume in the molten metal while cooling to the solidification point, (2) the decrease in volume while changing from the liquid to the solid state, and (3) the decrease in volume while cooling from the solidifying temperature to room temperature.

Pipes.

5. Pipes are cavities in the solidified metal caused by contraction and are formed during the change from the liquid to the solid state.

Contraction Stresses.

6. Contraction stresses are produced by hindered contraction and/or non-uniform cooling resulting in a non-uniform change in the volume of the metal. Contraction may be hindered by the resistance of the mold or by resistance within the casting itself due to non-uniform solidification.

Hot Tears.

7. Hot tears, or contraction cracks, are cracks in the metal formed at elevated temperature by contraction stresses. The steel is subjected to stress in excess of the tensile strength corresponding to the prevailing temperature and, accordingly, cracks. The formation of these cracks may be aided by contraction cavities (pipes) which cause a concentration of stress.

Cold Cracks.

8. Cold cracks are the result of large inherent contraction stresses produced during cooling from the solidification temperature to room temperature. These cracks appear at comparatively low temperatures.

SOLIDIFICATION

9. The process of the solidification of metals has been studied both from the theoretical and the practical point of view by many competent observers. We refer to H. M. Howe,^{1*} A. L. Field,² A. McCance,³ H. C. Dews,⁴ T. M. Service,⁵ C. H. Desch,⁶ J. E. Fletcher⁷ and others whose studies in solidification have formulated certain rather definite rules.

10. The solidification of steel poured into the mold begins with the formation of a crust. As the temperature drops the solidification proceeds and the crust extends inward until the entire cross-sectional area is solidified. This crust formation does not proceed uniformly all over the casting; at feeding and gating centers it proceeds more slowly.

11. Castings composed of thick and thin sections will solidify so that the thick portions are still in the semi-fluid or fluid state, while the thin parts of the casting are solidifying and may actually be accomplishing part of their solid contraction. Let us consider these three conditions, liquid, liquid plus solid, and solid, with respect to contraction.

12. The authors find that there are, apparently, no reliable published figures on the changes of the volume of liquid steel with change in temperature. However, from the information at hand it seems reasonable to assume that contraction takes place uniformly with the drop in temperature and that it is about 2 to 4 times greater per degree drop in temperature than solid contraction. Thus, in order to keep liquid contraction low, the steel should be run as cold as possible. Practically, this feature is not considered to be of importance as the supply of liquid metal usually is ample to provide for the amount of liquid contraction that might be obtained, regardless of the pouring temperature.

13. The freezing contraction is much more important and much more difficult to control. Experiments by Benedick⁸ and the calculations by Honda⁹ have shown that steel passing from the liquid phase to the solid phase contracts about 5.5 per cent in

* Reference numbers as shown herein correspond with numbers in the Bibliography at the end of the paper.

volume. The free solid contraction of steel, from the freezing point to room temperature, is about 6.6 per cent (linear contraction about 2.2 per cent). This type of contraction will be taken up more fully in the following pages.

14. T. M. Service⁶ explains the amount of contraction in steel in the following manner: Solid steel weighs 490 lbs. per cu. ft. (density 7.85) at 1500 degs. Cent. (2732 degs. Fahr.), while fluid steel weighs 437 lbs. per cu. ft. (density of 7.0). Therefore, in order to obtain one cubic foot of solid steel, 1.12 cu. ft. of fluid steel is required, or a total shrinkage of 10.72 per cent.

15. It has already been considered that in the solidification of steel castings the solidification proceeds from the surface inward and as it proceeds the three types of contraction take place simultaneously. Thus, when the casting has completely solidified there will be a contraction cavity at that place which was last to solidify, unless certain measures are taken to prevent its formation.

16. Pipes usually are thought of as a result of shrinkage alone, although some authorities still contend that occluded gases are also to be considered. This theory was first propounded by Fletcher⁷ as a result of his studies on ingot solidification. Other authors apparently do not consider the phenomenon of gas inclusion to be particularly important as a cause of pipes in steel castings.

17. The presence of pipes in castings may be considered in many ways. Heuvers¹¹ is of the opinion that pipes may be treated in three ways. First, the pipes may be ignored as, in his opinion, they do not affect the strength of the casting except insofar as they favor the formation of hot tears due to stress concentration at these points. Second, the pipes may be prevented by the insertion of metal chills at points of considerable cross-sectional area, or by the use of chill plates. Third, the design must be such that the foundryman can get to each section of the casting to feed it through direct feeding heads, or, at least, through sections of similar or larger cross-section.

18. These measures consist in casting in such a manner that the last solidifying portion will have an ample amount of metal to replace that lost during contraction. If the casting is not fed from some molten reservoir, a cavity will form in that portion of the casting that is last to solidify, or in any portion that is cut off from the supply of molten metal. In this case the volume of such cavities depends on (a) the mass of the metal solidifying, (b) the volume change of the metal, and (c) the casting tempera-

ture. Part (c) is really, for that matter, a portion of (b) because the extent to which casting temperature influences contraction is governed essentially by liquid contraction.

19. After a close analysis of the available literature on casting solidification, it is found that those who have expressed their ideas on the subject will generally end their discussion by pointing out that, if a casting could maintain an equal cooling rate and solidify uniformly throughout the casting there would be less defects from cracking of the casting. This could be done, they explain, by retarding the cooling of the thinner parts and by chilling the thicker parts to equalize the temperature of the casting.

20. It is true that if this can be accomplished, contraction stress arising from unequal cooling of sections would be eliminated. It is a nice picture, but it is fundamentally unsound for several reasons.

21. A casting with an equalized temperature throughout will proceed, upon reaching the solidification stage, to solidify from numerous points throughout. These points as nuclei for solidification will spread, and at those places where the solidified portions came together there will be small contraction cavities. It is true that there should be no large pipe or cavity but, instead, there would be minute cavities throughout the casting and a general unsoundness would be the result. Nor does such a method insure against the formation of cracks, as other investigators as well as the authors show that pipes and cavities are of importance in the formation of hot tears.

22. The casting cools from the face of the mold inward, as the mold acts as an effective chill. How, then, can cooling of thinner parts be retarded so that an equal temperature throughout the casting may be obtained? Even though the thick portions of the casting were internally chilled to bring them to the temperature of the thinner portions, the casting still would not solidify equally throughout but a skin of a certain thickness would always precede the mass of the casting in solidifying. A high-frequency coil might be used to replace the radiated heat by induction, but this practice would add greatly to expense.

23. Solidification by uniform cooling must be discarded and what has been called *directional solidification* offered in its place. Directional solidification is a process that is intimately involved with the design of the casting. It makes use of the idea that there should be a reservoir of liquid metal adjacent to the solidifying portions of the casting. As the casting proceeds in solidifying, the

last liquid portion is fed directly by a head so placed that it will provide the required liquid metal to compensate for the final solidifying contraction and thus promote complete metal soundness.

24. The outstanding principle in casting design, to obtain directional solidification, is to feed a smaller section through a heavier one. Of course, at times the design is such that this is impossible, and then the correct and intelligent use of chills may be resorted to in order that the heavier section will complete its solidification prior to the lighter section through which it must be fed. Directional solidification also necessitates the precaution that molds and cores should be made as collapsible as possible in order to reduce the stress arising from mold resistance.

25. A plan of solidification involving controlled directional feeding is especially adaptable as it employs the methods that are involved in the regular solidification of a steel casting. It does more—it does not permit sections to solidify heterogeneously throughout the casting, but controls the directions of solidification so that adequate reservoirs of liquid metal will feed the last solidifying portions.

26. The authors are of the opinion that in order to obtain castings free from pipes and voids it is necessary to have controlled directional solidification taking place within the casting.

Rate of Skin Formation.

27. A question that is ever prevalent when directional solidification is considered, and always asked by those interested in this method, is this: At what speed does directional solidification proceed? In other words, in a study of this nature the rate of skin formation is one of the important factors involved.

28. The rate of skin formation, based on our present information, depends on two conditions: First, the casting temperature, or perhaps better, the fluidity of the poured steel; and second, the rate at which heat can be conducted away from the mold-metal interface. Thus, the surface area and the concentration of mass of a casting would be influential in varying the degree of heat conductivity. In some sections a substantial skin is formed during the few seconds required for pouring the casting. In other larger sections where the concentration of mass is much greater, no skin at all will be formed during the pouring interval.

29. Constant volumes with varying surface areas will form, in a definite time interval, varying skin thicknesses. However, the

change is not critical, for a 50 per cent greater surface area does not change the rate of skin formation. To demonstrate this fact, an experiment was performed using a 6-inch diameter sphere and a parallelepiped, with volume identical to that of the sphere.

30. The surface area of the parallelepiped was 50 per cent greater than that of the sphere. Each was gated with a 2-inch central downgate. The metal was poured into the casting through the central downgate, and after a predetermined lapse of time the mold was turned over and the liquid metal remaining was emptied through the pouring gate. The pouring temperatures for the entire experiment were between 1550 and 1500 degs. Cent. (2822 and 2732 degs. Fahr.) The results obtained are shown in Table 1.

Table 1
EFFECT OF CONSTANT VOLUME AND VARYING SURFACE AREA
ON SKIN THICKNESS

Time elapsed between filling mold and over- turning mold, in minutes.	Linear thickness of solidified shell, in inches.	
	Sphere.	Parallelepiped.
0	0.18	0.18
1/2	0.31	0.32
1	0.56	0.55
2	0.75	0.75

31. Experiments involving the changing of surface area and the concentration of mass in regard to their effect on the rate of skin formation are being studied further by the authors. A study of temperature conditions and the rate of heat conductivity in the mold also is receiving attention. This is very important if the entire story on the rate of skin formation is to be known.

Fluidity Tests.

32. It has been mentioned that fluidity is a factor in the process of solidification. The difficulty is that, although it is recognized by all foundrymen as a dominant feature in casting technique, very little is known about controlling it. In fact, its control is one of the major problems in steelmaking today.

33. It has been known for some time that certain processes produce a more fluid steel at lower temperatures. We refer particularly to converter steel and point out that it is for this reason as much as any other that European foundries have maintained the converter practice. Just how much difference there is in the degree of fluidity of the best made open-hearth and electric, or acid

and basic steels, is not definitely known. However it is evident that the fluidity of steel is not altogether a function of temperature.

34. Which is to be considered the more important, temperature or fluidity? We are inclined to believe that fluidity is of greater importance, since the primary consideration of every foundryman is to run the casting completely. Then, too, the greater the fluidity, the easier it is to control directional solidification because the sections are more easily fed. The greater the fluidity, the greater opportunity there is, as will be shown later on, to fill up hot tears with liquid steel.

35. On a recent visit by Captain Shane, U. S. N., to the steel foundries of Europe it was discovered that a fluidity test was being used, especially by the French foundrymen, as an indication of the correct tapping or pouring conditions. These foundrymen claimed a fair correlation of conditions existed and that favorable results were obtained.

36. The use of a fluidity test instead of a stop-watch test seems to be the ideal way of obtaining advance knowledge of the mold-running ability. However, at the present time it appears that a definite correlation of conditions will be difficult to obtain; at least this has been the experience of the International Nickel Co. and the American Steel Foundries. At the Naval Research Laboratory, fluidity is being studied by using a modification of the Bureau of Standards horizontal spiral fluidity mold.²¹ The present data would indicate that numerous tests must be made before quantitative correlations can be obtained.

CONTRACTION AND TEMPERATURE OF HOT TEAR FORMATION

37. The casting of steel presents several difficulties which arise entirely from the design of the castings and the properties of the steel itself. The hot crack is one of the foundrymen's chief difficulties. Hot cracks and pipes are so closely associated with each other that it is difficult to differentiate between the two. It is the authors' opinion that many hot cracks are due in part to small pipes which serve as points of stress concentration and facilitate tearing in the weak steel.

38. Information concerning the critical temperature at which cracking takes place and the strength of the steel at that temperature has been lacking for sometime. Several investigations have recently been carried out to determine these facts, notably those by Körber and Schitzkowski.¹² A few other writers such as Heu-

vers,¹¹ Singer,¹⁴ Bennick¹⁴ and Osann¹⁴ also have contributed opinions on this interesting subject. In order to clarify the present views it might be well to mention some of these experiments and opinions in more detail.

39. Körber and Schitzkowski,¹² in what is probably the most comprehensive work of this character published to date, have determined the total shrinkage of cast steel to be approximately 2.18 per cent and the critical cracking temperature to be about 1300 degs. Cent. (2372 degs. Fahr.) The experiments carried out to determine these facts were made on three types of bars, in dry and green sand using both acid and basic open-hearth steel. The bars in question were (1) plain round bars, (2) bars with large flanges to hinder shrinkage, and (3) bars identical with those used in case (2) with the shrinkage hindered by the insertion of rigid iron bars in green-sand molds.

40. The amount of shrinkage was measured by the shrinkage meter developed by Wüst¹⁵ for his researches on cast iron. The bars were cast horizontally and the temperature measured by a platinum thermocouple, the hot junction of which was inserted at the surface of contact of the bar and the gate. The compositions of the steels used were as follows:

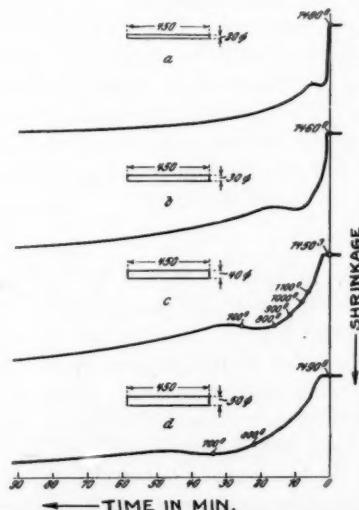


FIG. 1.—SHRINKAGE CHARACTERISTICS OF PLAIN ROUND BARS. (ACCORDING TO KÖRBER AND SCHITZKOWSKI.) TEMPERATURES EXPRESSED IN DEGS. CENT. MEASUREMENT OF BARS GIVEN IN MM.

	Basic Open-Hearth.	Acid Open-Hearth.
Carbon, %	0.15-0.39	0.28
Silicon, %	0.32-0.47	0.22-0.60
Manganese, %	0.49-0.80	0.26-0.69
Phosphorus, %	0.014-0.047	0.063-0.106
Sulphur, %	0.021-0.045	0.036-0.085

41. As the experiments with acid and basic steels were carried on in identically the same manner and the results proved conclusively that the shrinkage characteristics of the two types of steel were the same, only one set of the experiments will be given in detail.

42. The fact that acid and basic open-hearth steel have the same shrinkage characteristics is particularly interesting, since various opinions have been presented as to which type of steel is more conducive to cracking. None of these opinions, however, have been substantiated by actual measurements under identical conditions, and therefore they have not carried a great deal of weight.

43. The procedure followed and the results obtained are summarized below.

A—Plain Round Bars.

44. The bars used were 450 mm. (17.72 in.) long and 20 mm., 30 mm., 40 mm. and 50 mm. (0.787, 1.181, 1.575 and 1.969 in., respectively) in diameter. These bars were all cast in green sand and measurements taken on shrinkage. The curves obtained are shown in Fig. 1. The highest point on the curve represents the pearlite point. The total shrinkage was found to be independent of the diameter of the bars and was about 2.18 per cent. Approximately one-half of the total shrinkage occurred before the bars passed through the critical point.

B—Bars with Flanges.

45. The bars in this experiment were identical with those used above, except that they had flanges 20 mm. (0.787 in.) thick and 12, 13, 14 and 14 mm. (4.73, 5.11, 5.51, 5.51 in., respectively) in diameter. The molds in this case consisted of green-sand and dry-sand molds composed of compo (chamotte) fireclay and graphite. The curves *a*, *b*, *c* and *d* (Fig. 2) represent the data obtained on the bars cast in green sand, and *a'*, *b'*, *c'* and *d'*, the data obtained on the bars cast in dry-sand molds. The total shrinkage obtained when using green-sand molds, approximately 2.07 per cent, was found to be almost as great as that of a freely contracting plain round bar, with about one-half of the total amount again

occurring before the critical temperature was reached. These bars showed no defects.

46. In the curves *a'*, *b'*, *c'* and *d'*, representing the bars cast in dry sand, a deviation from the normal course is noted at about 1300 degs. Cent. (2372 degs. Fahr.) This deviation indicates a fracture in the bar; all of the bars fractured. The two larger bars fractured below the gate in the center of the bar, whereas both flanges were torn off the two smaller bars. None of these fractures showed any evidence of plastic deformation.

47. In this experiment the total shrinkage was found to increase with the increasing diameter of the bar and amounted to 0.34, 0.61, 0.89 and 1.78 per cent, respectively. The shrinkage before and after the critical point was 0.21, 0.18, 0.41, 0.89 per cent and 0.13, 0.41, 0.48, 0.89 per cent, respectively. In the two larger

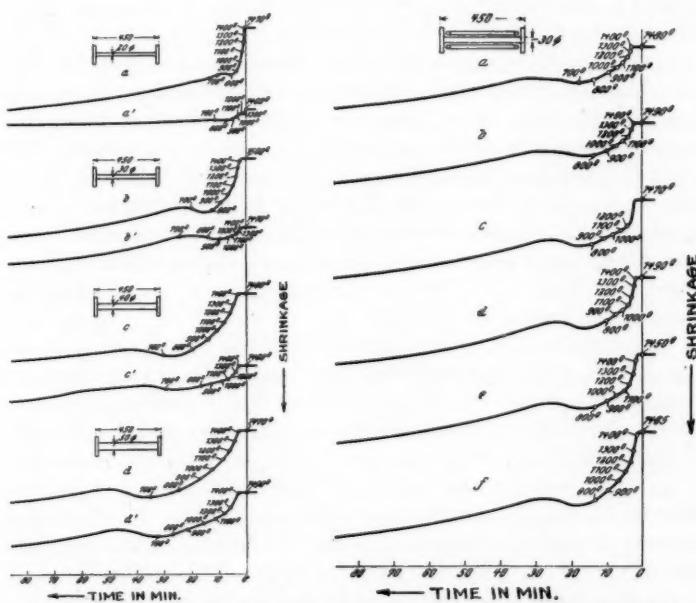


FIG. 2 (LEFT)—SHRINKAGE CHARACTERISTICS OF BARS WITH FLANGES. (ACCORDING TO KÖRBER AND SCHITZKOWSKI.) TEMPERATURES EXPRESSED IN DEGS. CENT.; MEASUREMENT OF BARS GIVEN IN MM.

FIG. 3 (RIGHT)—SHRINKAGE CHARACTERISTICS OF BAR WITH FLANGES AND RESTRAINING BARS IN THE MOLD. (ACCORDING TO KÖRBER AND SCHITZKOWSKI.) TEMPERATURES EXPRESSED IN DEGS. CENT.; MEASUREMENTS OF BAR GIVEN IN MM.

bars the shrinkage before and after the critical point is again approximately equal.

C—Bar with Flanges with Iron Rods in the Molds.

48. The experiment was conducted in green sand only and was intended to produce a mold the resistance of which was between that of a green-sand mold and a dry-sand mold. In this case only the 30 mm. (1.181 in.) bar was used. The shrinkage was hindered by rods of 20 mm. (0.787 in.) diameter placed between the flanges.

49. Various lengths of restraining rods were used to obtain different amounts of resistance. The total distances between the ends of the rods and the flange were 0.0787, 0.157, 0.236, 0.315, 0.397 and 0.472 in. In the two latter cases the distance was greater than the total shrinkage of the bar, 0.354 in., but the sand compressed between the end of the rod and the flange diminished the actual working distance somewhat.

50. The total shrinkage varied from 1.51 per cent for the most hindered contraction to 2.10 per cent for the bar that was impeded the least. The contraction after the critical point remained constant in every case, whereas the contraction before the critical varied from 0.51 per cent to 1.04 per cent. All of the bars were torn near the flange. No defect was visible in the three bars which had been impeded the least until about 0.039 in. had been turned off, when small cracks were found, connected to a small pipe in the flange. The shrinkage curves (Fig. 3) show deviations marking the fracturing points.

51. Although the work given above has presented the most complete data on the subject, and the conclusions as to the critical cracking temperature probably are very nearly correct, there are some portions of the technique that could be improved. The use of plain round bars with the gate in the center is not conducive to uniform cooling; that is, the cooling must proceed from the ends of the bar toward the gate. The ends of the bar thus pass through the critical range earlier than the center of the bar, and the shrinkage recorded is the summation of the expansion and contraction occurring throughout the entire bar. This would account for the "flattening" of the critical point, which should show as marked peaks on the curves. The temperatures recorded at the gate represent the hottest points and not the temperature of the bar as a whole.

52. In experiments 2 and 3, where the flanged bars are em-

ployed, similar difficulties are encountered. The thermocouple in this case does not indicate the "true" cracking temperature but some temperature slightly below it. This is brought out by the fact that the majority of cracks occurred at the flanges, which would indicate that the hotter and weaker metal was located there. The variation between the true temperature and that noted by the thermocouple at the center is not very great, due to the large amounts of heat given off at various points in the bar as these points pass through the critical range, which tends to equalize the temperature to a great extent.

53. The authors are at the present time conducting a series of experiments in an endeavor to continue the work of Körber and Schitzkowski and determine the tensile strength of steel at the high temperatures where cracking takes place. Although the data are at present incomplete and not sufficient to warrant a conclusive statement at this time, we are of the opinion that the temperatures given above are very nearly correct.

54. In order to correct the experimental errors mentioned above, the authors made a study of the cooling characteristics of several bars. Three platinum platinum-rhodium and six iron-chromel thermocouples were used in noting the temperature in various portions of the bar. The three platinum couples were placed as follows: (1) about one inch from the end of the bar, the end of the thermocouple extending to the center line of the bar; (2) at the center of the bar, below the gate; (3) about three inches from

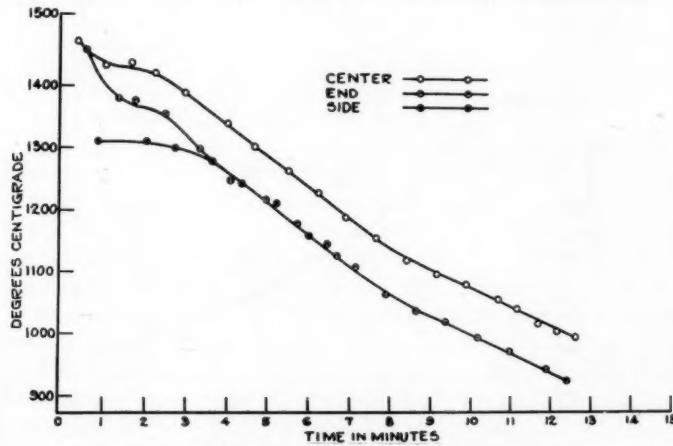


FIG. 4—COOLING CURVES OF A 2-INCH ROUND BAR.

the end of the bar, with the end of the thermocouple protruding only about one-eighth inch into the mold so that the temperature of the skin would be recorded.

55. In Fig. 4 are shown the data obtained on a straight bar 2 inches in diameter. This bar was similar to that used by Körber and Schitzkowski in a portion of their work. It will be noted that the skin cools immediately to almost 1300 degs. Cent. (2372 degs. Fahr.) and then cools very slowly. The fact that solidification proceeds from the end toward the center is shown very markedly by the fact that the temperature just below the gate lags about 75 degs. Cent. (135 degs. Fahr.) behind that of the end and the skin in cooling.

56. To correct this, a bar has been selected having a diameter at the center slightly smaller than that at the ends. The thermal characteristics of this bar are shown in Fig. 5. The thermocouples were placed in relative positions so that the data are comparable. It will be noted that the cooling is almost uniform throughout the bar.

57. Experiments to determine the maximum shrinkage and the critical cracking temperature of steel have also been conducted by Heuvers. His results, which he presents with no explanation of the technique used, agree very well with those given above and show that one-half of the total shrinkage of 2 per cent occurs within the limits of 1450 and 900 degs. Cent. (2642 and 1652 degs. Fahr.) within a very short time, and that below 900 degs. Cent. the additional shrinkage is very slow. He states further that

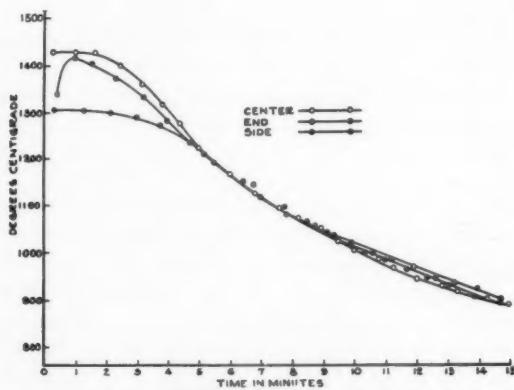


FIG. 5—COOLING CURVES OF A DOUBLE CONE-SHAPED BAR.

"within the limits mentioned, *i. e.*, during the transition from the fluid to the pasty state and to the state of white-hot heat, the steel is rapidly subjected to great stress at a time when the tensile strength is still slight," and contends that these facts corroborate the work of Körber and Schitzkowski.

58. K. Singer^{13, 14} and Krieger,¹⁴ who have also studied the problem, are of the opinion that the results presented by Körber and Schitzkowski are very nearly correct. The only published statement that differs considerably from those given above was presented by B. Ossan in a discussion of Heuvers' paper.¹⁴ It is Ossan's opinion that the critical cracking temperature is very much below that given above. He believes that the true temperature is approximately 600 degs. Cent. (1112 degs. Fahr.), and bases his argument upon the following points:

That it is impossible to determine the temperature of the cracked skin by means of a thermocouple placed in the mold.

"That a crack cannot occur as long as the steel is still in the fluid and, consequently, in a plastic state, because stress in this state causes deformation but no cracks. The plastic state does not cease until a temperature which is about 600 degs. Cent. (1112 degs. Fahr.) is reached. Therefore, cracks can occur only when the steel has cooled off to that temperature, *i. e.*, when the skin has reached this temperature. However, the thermocouple introduced into the mold may still indicate 1300 degs. Cent. (2372 degs. Fahr.)"

"That the cooling of the steel along the mold wall proceeds very rapidly due to the great temperature gradient. The critical temperature of 1300 degs. Cent. . . will be reached in such a short time that it would be impossible to 'withdraw the mold.' "

59. Ossan continues that such wide temperature differences as he has mentioned, skin at 600 degs. Cent. (1112 degs. Fahr.) while the center is still at or above 1300 degs. Cent. (2372 degs. Fahr.), do exist. As an example he points out that ingots, whose external appearance is quite dark, begin to glow white hot when placed in the "soaking pit" due to the molten state of the interior.

60. In presenting this hypothesis Ossan has neglected to take into account several points of importance. First, an ingot usually is cast in a metal mold, which not only has a much greater chilling effect but also has a much higher heat conductivity than any sand mold. Second, in removing the ingot to the "soaking pit" it must of necessity be exposed to the chilling effect of the air, which,

again, is much greater than that experienced by a casting in a sand mold.

61. The supposition that such wide temperature differences can exist in a casting produced in a sand mold becomes untenable when it is considered that the steel is poured into the mold at a temperature which is above the melting point, and that before crystallization can begin in the outermost crust the heat released during cooling of the steel to the point of solidification and also the heat of crystallization during solidification must be absorbed. As the latter quantity of heat represents approximately one-sixth of the total heat content of the fluid steel, it can readily be seen that the mold becomes very hot and that, due to the poor thermal conductivity of the sand, cooling of the casting cannot proceed at the rapid rate suggested.

62. These arguments are more or less confirmed by the results of the temperature measurements made by F. Wüst and P. Stuhlen¹⁶ in their studies of temperature differences at different points of gray iron bars during the cooling process. They found that up to the pearlite point no temperature differences much over 100 degs. Cent. (212 degs. Fahr.) occur between different points of the bar. This can also be seen by examining the curves shown above in the discussion of the work of Körber and Schitzkowski.

63. The authors, in an endeavor to determine how much of a temperature gradient existed in a sand mold, took temperature

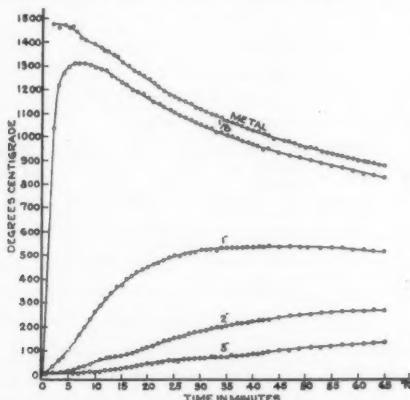


FIG. 6—TEMPERATURE CURVES SHOWING RELATION BETWEEN A 6-INCH CAST STEEL SPHERE AND SURROUNDING SAND; 1/8 IN., 1 IN., 2 IN. AND 3 IN. FROM THE CASTING SURFACE.

measurements on a mold in which a 6-inch diameter steel sphere was cast. The temperature of the metal at the center of the sphere was determined by means of a platinum platinum-rhodium thermocouple protected by a quartz tube, introduced through the bottom of the mold. The temperature of the sand was noted by means of four chromel-alumel thermocouples placed 1/8, 1, 2 and 3 in. from the mold-metal interface. Potentiometer readings were taken at short intervals and plotted, as shown in Fig. 6.

64. By examining these curves it can readily be seen that the opinion presented by Ossan is untenable. The thermal conductivity of sand is so low that even though the temperature of the sand at the face of the casting is only 150 degrees lower than the metal at the center during almost the entire cooling range, the sand temperature one inch away from the casting never exceeds 1000 degs. Cent. (1832 degs. Fahr.), and the sand two and three inches away is still lower in temperature.

65. The results of Körber and Schitzkowski may also be cited as proof that the cracking temperature chosen by Ossan is too low. In their experiments it was found that the cracking occurred where the highest temperatures prevailed, and that if shrinkage was not prevented until the temperature of this point in the bar had dropped to below 1000 degs. Cent. (1832 degs. Fahr.), no cracks appeared. It has also been pointed out that in the region of 600 degs. Cent. (1112 degs. Fahr.) and less, steel has such a high tenacity that cast steel bars in this temperature region break only after previous plastic deformation.

66. This is true even in the case of notch impact tests, which, because of the very nature of the stress, favor the formation of brittle fractures.¹⁷ Therefore, if the cracks observed in the test bars at the transition to the flanges, or the gate, had occurred at the low temperature of 600 degs. Cent. (1112 degs. Fahr.), preliminary plastic deformation might be expected.

67. No such plastic deformation was reported. On the contrary, it was found that the cracks always exhibited the characteristics of a fracture without plastic deformation. These fractures are explained by Körber and Schitzkowski by the assumption that slightly below the temperature at which solidification sets in, between the crystals which separate first and which form the first solid crust, small amounts of mother liquor melting at a lower temperature are still present.

68. K. Singer is another proponent of the higher cracking

temperature. He, however, uses his theory of solidification, the refilling of cracks by molten metal, to explain why cracks are not more frequently found. His theory will be treated in detail in subsequent pages.

FACTORS EFFECTING THE FORMATION OF HOT TEARS

69. Undoubtedly there are a number of factors effecting the formation of hot tears in steel castings. The foundryman's knowledge on this subject has been limited by the lack of fundamental research, and the result is that certain variables can be spoken of in only a qualitative way. It is very probable that some controlling factors which future work will stress more completely may be entirely missing from present discussions of this subject.

70. The factors that are at present deemed most important in the formation of hot tears are as follows:

- (A) Contraction stresses—
 - (1) Caused by mold resistance.
 - (2) Caused by uncontrolled directional cooling.
- (B) Pipes or cavities.
- (C) The physical properties of the steel.
 - (1) Strength at solidification temperatures.
 - (2) Fluidity.

Contraction Stresses.

71. Mold resistance is very conducive to the formation of hot tears. It has already been pointed out in the review of the work of Körber and Schitzkowski that mold resistance can be sufficiently great as to fracture bars. It is hard to realize that rammed sand can exert such a resistance to the contraction of steel that fractures result.

72. The fact that some sands have a high compression strength and resist the contraction of the steel is not the entire story. The mold hardness and the amount of sand around the casting also are important.

73. Take, for example, the H-shaped section such as is often encountered in pulley wheels, gear blanks, valve bodies and the like. The sand between the two flanges restricts the casting from moving. It is true that the sand lying next to the casting loses its strength as it becomes burnt and that it will allow for movement of the two flanges toward each other, but the casting has little opportunity to avail itself of this small amount of play because the rest of the sand mold is rigidly shaped. Thus, it is not only a

question of overcoming the compression strength of the sand between the flanges, but it is also a case of overcoming the resistance of the entire mold to the casting movement.

74. It is perhaps well known that green sand molds have less mold resistance than dry sand molds for the reason that dry sand molds will have a greater hardness and the sand a greater compression strength. Castings that crack in dry sand may not crack at all in green sand.

75. Cores are especially responsible for mold resistance. The core irons and reinforcing bars offer a resistance that is difficult to overcome. Later on in this article, methods will be considered, whereby mold resistance can be materially decreased and in many cases eliminated.

76. The metal itself can set up contraction stresses of such a nature that cracking takes place. These stresses are the direct result of the manner in which the casting solidifies. The concentration of these stresses usually takes place at a rapid change in section of the casting where the thinner portion is solidified and the heavier portion partly fluid. The thinner portion would be contracting and the concentration of these stresses will be at the weakest point, which is the junction of the thin section with the thick section.

77. In many cases the stresses are concentrated by the solidifying metal and, as the mold resistance prevents contraction of the casting, hot tears are formed. It has been claimed that the concentration of these stresses is caused by uncontrolled directional solidification. If no thought has been given to the rates of solidification of various sections of the casting, the metal will solidify in such a heterogeneous manner that a thin section which has totally solidified will pull away from the thicker section which is in the critical range for crack formation.

Pipes or Cavities.

78. The authors consider that pipes and small cavities are factors in the formation of hot tears as they become centers of stress concentration. These views are not held by everyone. K. Singer¹³ is of the opinion that cracking is not associated with the formation of pipes and that the crack forms earlier than the pipe.

79. It is hard to believe that this statement tells the entire story about the formation of cracks. It is quite possible that cracks do occur while the casting is solidifying, but it is equally possible that cracking may occur after solidification is complete.

and while the casting is at high temperatures—for example, between 1300 degs. Cent. (2372 degs. Fahr.) and the solidifying temperature.

80. Pipes and cavities form at the last point of solidification of a section, and thus are inherently the weakest points of the casting. It is well known that all types of cavities are points of stress centralization, and thus the two factors of metal weakness and stress concentration produce hot-tear cracks.

81. Theorizing is all very well, but are hot tears and cavities found to exist simultaneously in practice? The radiograph of Fig. 7 and the photograph of Fig. 8 show one of the numerous cases the authors have found in studying defective castings.

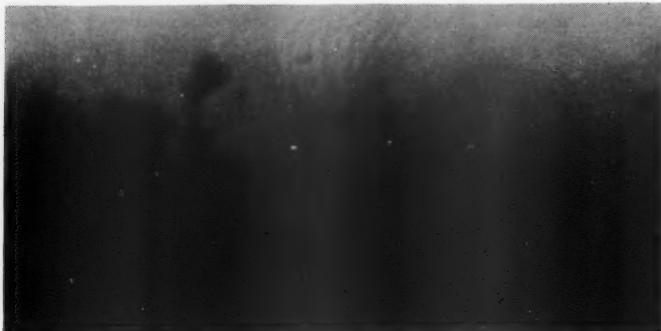


FIG. 7—GAMMA RAY RADIOPHOTOGRAPH SHOWING PRESENCE OF A HOT TEAR EXTENDING FROM A CAVITY.



FIG. 8—HOT TEAR EXTENDING FROM A CAVITY IN CAST STEEL SECTION.

82. The radiograph (Fig. 7) shows hot-tear cracks leading from a cavity. The photograph (Fig. 8) shows another section of a casting that has been machined. It will be noticed that the cavity is in the center of a change of casting section where an abrupt change is made to a much thicker section. The cracks are clearly seen leading out of the cavity.

83. It has already been pointed out that abrupt changes in section are centers for stress centralization, and that if there is a lack of controlled directional cooling, pipes will form which become the point of stress application. The result is, as we see in the photograph, hot-tear cracks. The photographs also offer the suggestions that cracking takes place after solidification, or perhaps during the mushy stage of solidification, as otherwise the crack would not have been found penetrating the metal surrounding the cavity, since this portion of the crack would then have been filled by the last solidifying metal.

84. It frequently occurs that an external tear is indicative of the presence of an enclosed pipe or shrinkage cavity, particularly at the radiused junction of a flange with a body.

Physical Properties of the Steel.

85. The information that is available on the physical properties of steel at temperatures closely approaching the solidification temperature are nothing more than theories or suppositions resulting from an attempted interpretation of conditions that have been found by experience to exist. The greatest difficulty in this regard is that there are so many contradictory opinions that it must be admitted that our knowledge is very scanty.

86. It can be seen that the formation of hot tears depends somewhat on the physical properties of the metal. It is true that the strength at these high temperatures is quite low—so low, in fact, that a slight change may increase the strength by 50 per cent, 100 per cent, or more.

87. Then, again, how important is ductility at these temperatures? Is it not possible that the metal elongates under stress at these high temperatures, and, if so, would it be possible to increase the ductility? These are merely a few of the large number of questions that need answering.

88. On this question of physical properties for resisting cracking, there are two paramount and somewhat contradictory ideas in vogue. It is held by certain practical foundrymen that the higher the carbon (preferably 0.35-0.40 per cent), the greater

the strength of the steel and therefore the casting is more able to withstand hot-tear formation. Another group holds that the lower the carbon (preferably 0.20-0.25 per cent), the greater the ductility of the steel at the cracking temperature and the casting is then able to withstand hot-tear formation due to stretching or elongation.

89. Which of these views is correct, or are they both correct? Perhaps the carbon content is only indirectly responsible for the effects that are noticed. Carbon is not the only outstanding example of the use of alloying elements to resist cracking.

90. It is claimed that manganese adds strength to the metal at temperatures just below its solidifying range, and that nickel toughens the metal so that it will elongate instead of fracturing. Perhaps there are other elements that may be alloyed that would increase the strength or ductility more than carbon, nickel or manganese.

91. Then, too, it is well to point out that carbon, nickel, manganese or any other alloying element may have no appreciable effect at all on the strength or ductility of cast steel in the temperature regions just below the solidification range. These features are important and should be studied with the idea of obtaining quantitative information. Attention should also be called to the important role that oxygen and oxides may play in controlling the physical properties at these high temperatures.

92. For the sake of clarity it is pointed out that this discussion of alloying elements would apply only to solidified steel. It is rather difficult to understand how alloying elements could bequeath strength or ductility properties to steel in the mushy state of the solidification range. As cracks do exist at this stage, other plans for their elimination must be formulated.

93. The Naval Research Laboratory at present has set up a research of this nature and, although the problem is a tedious one, it is expected that in the near future information on the strength of steel as cast at high temperatures will become available.

94. Before leaving the subject of cracking in completely solidified steel, it should be pointed out that the temperature at which the hot tears take place will depend on the strength of the metal and the amount of stress on the casting. The strength of the metal increases very gradually from no strength at all just prior to solidification to what is still a very low figure at 1300 degs. Cent. (2372 degs. Fahr.). Rosenhain³ has claimed that mild steel of the type used in steel castings has, in the neighborhood of 1100 degs.

Cent. (2012 degs. Fahr.), a strength of a little over 2000 lbs. per sq. in., while in the neighborhood of 1300 degs. Cent. (2372 degs. Fahr.) it is less than 800 lbs. per sq. in.

95. The amount of stress on the casting is that resulting from hindered contraction. Just prior to solidification of the casting the stress would be zero, and as the temperature drops the stress becomes greater. No data are available as to the comparable rates of increase in strength or stress due to restricted contraction at high temperatures. It has been observed, however, that in the neighborhood of 1300 degs. Cent. (2372 degs. Fahr.) the stresses developed by resisted contraction are sufficient to crack the casting. Below this temperature the strength of the steel increases more rapidly and is therefore sufficient to withstand the stresses that are developed by hindered contraction.

Fluidity.

96. French foundrymen have emphasized the importance of obtaining fluidity in cast steel, as in their opinion the smaller percentage of defects will be found in castings poured with a more fluid steel. There are points in favor of such a supposition. In the first place, the greater fluidity allows for better controlled directional solidification and the possibility of filling up cracks. If a crack should form in a portion of the casting already solidified and the crack extends to liquid metal, then the greater the fluidity of the steel, the more opportunity there is for the crack to become completely filled.

97. It is quite possible that fluidity is the controlling factor rather than strength, and that the higher carbon steels (0.35 per cent) are not so susceptible to cracking. Singer makes the statement: "The higher carbon steels remain fluid in the mold for a longer time and pieces cast from these steels are, as we know, less liable to crack."

98. Singer stresses the fact that fluidity is responsible for non-cracking. However, he has forgotten that the cracking may take place after the casting has completely solidified but while it is still at a very high temperature. Fluidity certainly would have no influence on such conditions.

TYPES OF HOT-TEAR CRACKS

99. In the preceding discussion the factors that are deemed most important in the formation, or in restricting the formation, of hot tears have been listed. It is now our plan to examine the

methods of formation of hot tears. It is the authors' opinion that there are two types of hot tear cracks, as follows:

(a) The internal hot tear that develops during the solidification of the casting, and is caused by a lack of controlled directional solidification.

(b) The external hot tear that develops after solidification is complete, and is caused by stresses due to mold resistance. These cracks occur at temperatures slightly below the solidification temperature.

The Internal Hot Tear.

100. Internal hot tears are of a decidedly ragged nature with numerous branches. They have no definite line of continuity and usually exist in groups. These tears seldom terminate at the surface; in fact, most of them do not. When they do appear at the surface they are usually very small and difficult to locate. Upon exploring them downward they will be found to branch out and become more pronounced. The radiographs of Figs. 9 and 10 show types of internal hot tears.

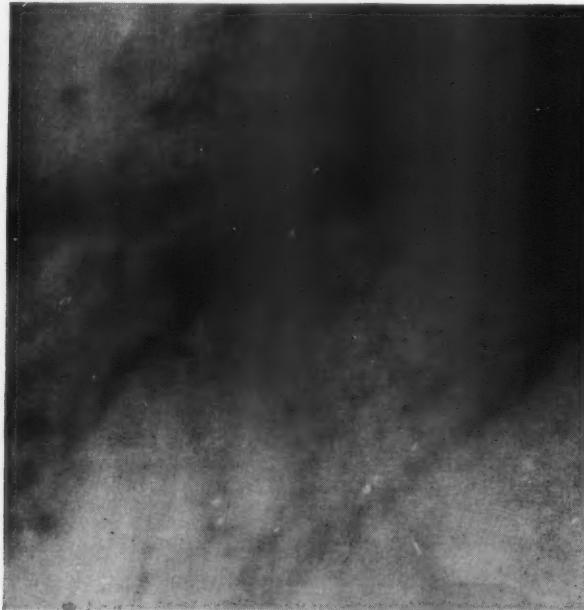


FIG. 9—GAMMA RAY RADIOGRAPH OF TYPICAL INTERNAL HOT TEAR.



FIG. 10—GAMMA RAY RADIOPHOTOGRAPH OF INTERNAL HOT TEAR FORMATION.

101. It has been said that the cause of the formation of these cracks can be traced directly to a lack of controlled directional solidification. It can readily be seen that the tearing is caused not only by the great differences of temperature in the casting, but that it also depends on the nature of the transition from the hottest to the coldest parts of the casting. A solidified section will pull away from a larger semi-solidified section if the transition between the sections is abrupt. However, if the transition is more gradual, the contraction forces are not so concentrated because solidification proceeds in a more controlled directional manner.

102. A lack of controlled directional solidification is also responsible for the lack of metal, and a cavity is the result. This cavity may form as a crack instead of a pipe in certain heavy sections, since the contraction of the metal toward the solidifying centers is sufficient to create a stress that tears the casting.

103. These tears occur and proceed from the inner solidifying face outward toward the surface, since the forces of contraction concentrated on the section will be sufficient to tear the weaker, less rigid portion of metal. Once the tear has started, it easily penetrates the stronger metal. This hypothesis can be correlated with the observed facts, as it is found that in the interior of the casting the tears are quite large, while they appear only as small cracks at the surface.

104. Enerustations and noticeable risings on the surface of castings have led to the opinion that crevices have formed in the steel casting which afterwards have been filled. Such enerustations of the surface have been found in castings at places where there is a danger of cracking, usually without any cracks in the surface being observed.

105. Singer and Bennek¹⁸ have very logically explained these observed conditions. It is their opinion that as the walls of the mold were perfectly smooth at these places before the piece was cast, the encrustation must have arisen after the shaping of the casting, presumably because the solidified outer skin of the casting was ruptured at this place, a crack being also produced in the wall of the mold adjacent and adhering to this outer skin, in the same direction as the crack in the casting. The fluid steel making its way into the crevice soon after the rupturing of the outer skin also filled the crevice in the mold and thus led to the encrustation.

106. The filling up of cracks is a subject on which there is little information, although the idea is not a novel one. The principles involved are very similar to conditions of rock formation and the actions of intrusives, but certain favorable conditions must prevail.

107. In the first place, the crack must open up from the inner solidifying wall of the casting and proceed outward to the surface. The steel section must be thick enough to allow for gradual solidification toward the center of the section, and the steel must be of such a degree of fluidity that it will be able to flow through the crack to the surface of the casting. Of course, the actual distance through which it flows may be very short, as the skin of the casting may be only a few millimeters thick when the crack takes place. The difficulty becomes more pronounced as the solidification proceeds.

108. It is the authors' opinion that the filling up of cracks is the exceptional rather than the usual case, for two reasons. The contraction stresses are not usually sufficient to crack the skin at the first phase of solidification when the liquid metal does not have to proceed far to fill them, and the fluidity of the steel is so low that the liquid steel could not traverse the crack when the stress became sufficient to produce a hot tear. The latter statement is based on the typical American practice of pouring heavy castings at as low a temperature as possible.

109. To proceed with the supposition that cracks may be occasionally filled, it is suggested by Singer and Bennek that the cracks may be filled either by the general solidifying metal or by the mother liquor. In the first case the metal is forced into the crack under the pressure of the fluid column of the pouring gate, and the crack as filled will not differ in chemical composition from the material surrounding it. In the second case the impure, later-solidifying mother liquor is forced into the crack under the pres-

sure of the contracting crust, and the crack in filling will have a higher carbon, phosphorus and sulphur content than the material adjacent to it.

110. Furthermore, they uphold their contention by exhibiting specimens of filled cracks which show segregations and, in places, a plainly visible transcrystallization of the filling metal in a direction perpendicular to the walls of the crack. If a crack can be filled with mother liquor in the manner described, then a filling with liquid reservoir metal is equally conceivable. It is rather difficult, however, to produce proof of this fact, since the material of the fractured outer skin and that filling up the crack usually shows no noteworthy differences in chemical composition.

111. The time at which the cracks occur in the first formed skin of a casting depends on the forces of contraction arising in the piece as it cools, hence, on the form of the piece. If the form is such that cracks occur in the outer skin while it is still thin, then these cracks may have an opportunity to be filled from the liquid reservoir, providing the steel is fluid enough.

The External Hot Tear.

112. The external hot tear develops after solidification is complete but while the casting is at a temperature slightly below the solidification temperature, and it is caused by stresses due to mold resistance. The reason that the crack has been termed external hot tear, is that the crack first opens on the surface of the casting and proceeds inward. Such a crack has the appearance of a tear; that is, the face of the crack is ragged. The crack is wider at the surface and is uninterrupted in depth.

113. The cracking or the failure of such a casting would be comparable to the failure of a test specimen under load at a high

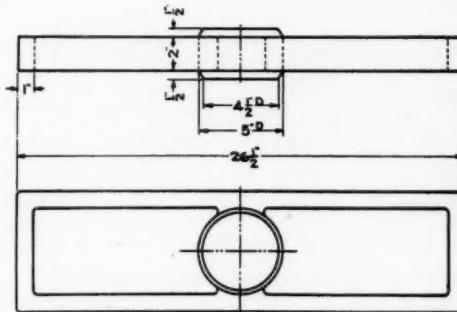


FIG. 11—PATTERN USED IN THE STUDY OF EXTERNAL HOT TEARS.

temperature; the cracking begins at surface irregularities which serve as points of stress concentration. It is obvious that mold resistance will depend upon the shape of the casting and the conditions of the molding sand. A mold that is collapsible in the danger zone can render the detrimental forces of contraction less effective.

114. In the study of external hot cracks, a pattern involving a hub and arm as shown in Fig. 11 was molded in 3 ways, (1) in dry sand, (2) in green sand, and (3) in green sand made more collapsible by hollowing out that portion between the arms. The radiographs of Fig. 12 show the amount of cracking as obtained by the variation of mold resistance. It will also be noticed that the cracks are wide at the surface and become smaller as they penetrate the metal.

115. The casting made in the collapsible mold shows no

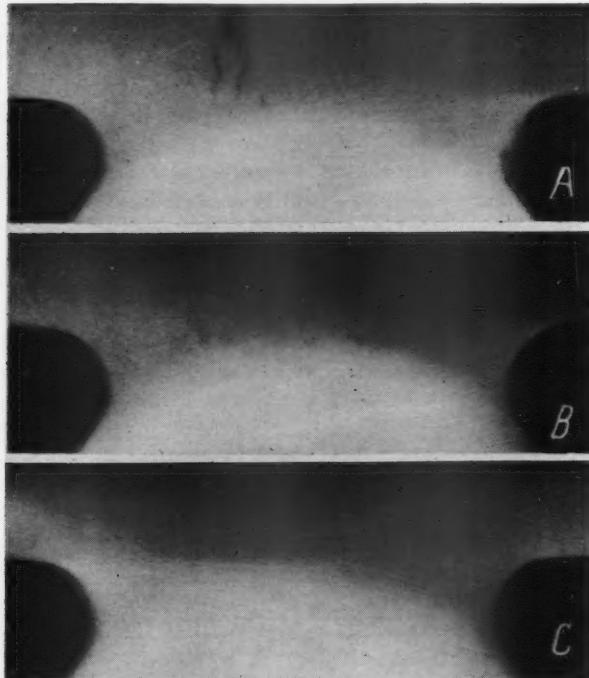


FIG. 12—EXTERNAL HOT TEARS AS OBTAINED BY VARIATION OF MOLD RESISTANCE. A: CASTING MADE IN DRY SAND. B: CASTING MADE IN GREEN SAND. C: CASTING PRODUCED IN COLLAPSSIBLE GREEN SAND MOLD.

cracks at all. It may then be assumed that mold resistance caused the cracks, and that the greater the mold resistance, the larger the crack. There is little possibility that these cracks formed before solidification was complete, since there is no evidence of large internal tears.

MEASURES EMPLOYED TO OBLVIAE HOT TEARS

116. The formation and development of hot tears having been studied, it appears wise to consider methods that may be adopted to obviate the danger of hot tears, or at least to limit the tendency of hot-tear formation. In this regard the following measures are employed:

- (a) Flexible molds and cores.
- (b) The use of chills at the proper places.
- (c) The application of ribs and wedges at certain changes of cross section.
- (d) The use of a thermal non-conductive molding material.
- (e) Casting in preheated molds.
- (f) Pouring from elevated temperatures.
- (g) Casting design.

(a) Collapsible Molds and Cores.

117. It has been stated already that flexible molds and cores have a beneficial effect. Considerable time must be used in producing such molds, and the sand recovery after molding often is very difficult when cinders and the like are used in backing up parts of the mold and in the center of cores. The use of such molds, of course, is limited.

118. The data concerning the critical cracking temperatures show how important it is to keep molds and cores collapsible so that their resistance does not cause cracks. In many cases, in order to prevent cracking, the mold and cores must be destroyed immediately after casting. Immediate action is necessary because, according to the experiments of Körber and Schitzkowski and our own data in the case of thicknesses of from 20 to 50 mm., a period of only 2 to 4 minutes intervenes between filling the casting and reaching the critical cracking temperature in the neighborhood of 1300 degs. Cent. (2372 degs. Fahr.)

119. Mold resistance arising from the negligent use of core irons has been repeatedly pointed out by G. Batty.¹⁹ He claims that utmost care should be used to see that core irons are straight and pointed on the ends so that they will not interfere with the collapsibility of the core.

120. Great pains are taken in European plants to obtain mold collapsibility by the use of relieving blocks, cinders and the like. All tie rods and reinforcing bars are placed in such a manner that they protrude from the ends of the flask. This prevents the mold resistance that would arise from the sand compressing between the ends of the bars and the flask.

(b) *The Use of Chills.*

121. The use of chills is always a very uncertain measure and their presence is likely to cause cracks, the very thing that it is so desirable to eliminate. Theoretically, chills cool portions of the casting that otherwise would remain fluid longer than the surrounding portions through which the larger section is fed. If chills of the correct size are used, the process of directional solidification would be aided, with the result that the formation of hot tears would be unlikely. In other words, the chill acts as a method for reducing the dimensions of the piece, and the dimensions of the piece as well as the fluidity of the steel at the time of pouring determine how long contraction and solidification continue.

122. To overcome poor design—it is really only because of poor design that chills are used at all—is rather difficult. In the first place, the amount of cooling should be known, and the answer at best is only a guess. Thus, the mass of the chill is merely estimated. If the chill is estimated as a greater mass than is really required, then the section is over-chilled and contraction stresses are set up in a manner opposite to that which would have been experienced if no chill were used, with the possibility of the formation of a hot tear.

123. It has been observed that the use of too large a chill plate may crack a casting, due to a too-rapid development of shrinkage stresses. Likewise, a center of solidification is started which in many cases may not be desired. If the mass of the chill is less than that required, then solidification will not proceed as rapidly as is necessary to insure a sound section.

124. The discussion above has applied, in particular, to external chills. The use of internal chills is even more critical. If the mass of an internal chill is too large it will not fuse in, and along with the detrimental features just pointed out there will be the added one of lack of cohesion of the metal. Internal chills should be clean and dry, otherwise porosity is apt to result.

125. It is the authors' opinion that internal chills are more conducive to the formation of hot tears and other internal defects

than they are to obviate these dangers. Their use is not scientifically controlled, and the dangers arising from their misuse are many.

126. Whatever can be accomplished by chills may be attained by the use of external chills with less danger of harmful results. It is true, of course, that the use of external chills involves more expense, since usually it is necessary to cast them in advance in order that they will fit the section which is to be chilled.

(c) *Application of Ribs (Brackets).*

127. The application of ribs has served to check hot-tear formations. They have been used rather extensively in this country with fairly good results. Their function is to solidify earlier than the sections which they adjoin, and in this manner act as a bond in preventing cracking.

128. Ribs do not eliminate the contraction stresses; they merely tend to prevent these stresses from centralizing at one point and rupturing the casting. The strength added to the surface skin by the presence of ribs acts to transfer converging stresses over a wide area.

129. With some castings even a generous supply of ribs will be of no use; although the ribs themselves remain solid, cracks occur between them. In such cases the designer must come to the aid of the foundryman by providing for gradual transitions and smaller junctions between the sections of varying thickness.

130. The use of wedged-shape attachments to the casting on which gates and risers rest often are advisable, because all measures that are taken to prevent the formation of pipes in the casting will effectively lessen the danger of hot-tear formation. A disadvantage in the use of ribs and wedge-shaped attachments is their removal cost, which sometimes is considerable.

131. A few other points should be considered in the prevention of hot tear-formation. It has been previously shown that contraction stresses that arise suddenly within a casting are most conducive to hot-tear formation. In order to take steps to prevent these stresses from arising so suddenly, a slower rate of solidification must be effected. Even if this is not entirely possible, any step that can be taken in this direction must have a good effect.

132. Of course, the first suggestion is to increase the wall thickness of the casting, for the greater the section, the slower the cooling. However, in some cases this is not altogether desirable

or feasible. The three other alternatives that are available are the use of thermal non-conductive molding material, casting in preheated molds, and pouring from elevated temperatures.

(d) *The Use of Thermal Non-Conductive Molding Material.*

133. It has been noticed that a skin forms quickly when molten steel is poured into a mold. The thickness or degree of formation of the skin depends largely upon the rate with which the mold material can conduct the heat away from the solidifying casting. Thus, the chilling action of the material of the mold has a great effect upon the rate of solidification of the steel.

134. If the mold material could be made less conductive of heat, the chilling action would be somewhat lessened. It is obvious that sand is much better in this regard than is a metal mold. A molding material that would have poor conducting properties must also have the good qualities that sand possesses, namely, adequate refractoriness and good moldability. These properties narrow the field perceptibly.

135. There remains magnesium oxide in one form or another, lime (plaster of Paris), asbestos and diatomaceous earth, and even these have various objectionable features. Asbestos and plaster of Paris lack good permeability, magnesium oxide and diatomaceous earth have little strength, and diatomaceous earth has such a high porosity that it would give trouble in casting because the air enclosed in the pores must escape in consequence of its expansion upon being heated. This air may attempt to escape through the steel, which would cause a porous casting.

136. It seems that, for one reason or another, these materials have limitations. However, they can be used in a mold, especially if they are used in combinations such as asbestos shreds and magnesium oxide or diatomaceous earth. A great moldability is thereby obtained and the heat conductivity considerably lowered. If these poor heat conductors are placed at the thin portions of the mold and the better conducting materials used at the thicker portions, a more even directional solidification may be obtained.

(e) *Casting in Preheated Molds.*

137. In regard to casting in preheated molds, it may be said that this is an excellent way to get around the excessive chilling action of the mold material. In a way, the same object is accomplished as in the use of thermal non-conducting materials, since the molten steel coming in contact with the warm walls of the mold does not cool so rapidly.

138. The preheating cannot be carried on at high temperatures, since the mold will crumble from lack of bond strength; then, too, the process is not favorably received because it causes difficulties in manufacturing schedules.

(f) *Pouring from Elevated Temperatures.*

139. Pouring from elevated temperatures as a means to avoid hot tears probably sounds like a foolish suggestion to most experienced steel foundrymen whose slogan for years has been "Pour her cold and prevent cracking." This opinion is urgently in need of correction. Steel poured hot will give excellent castings, and if it is a well-made, well-refined steel, thoroughly deoxidized, it will not have as great a tendency to the formation of hot tears as that which is poured cold. A great number of steel casting manufacturers do not make steel for castings but merely melt steel for casting; consequently, the formation of the slogan "Pour her cold."

140. It should be understood that the molds used in conjunction with pouring from elevated temperatures must be made of material which (a) will not readily be eroded by the high temperature metal, and (b) will not materially resist the progress of normal solid contraction of the casting. The gating and heading should be so devised as to insure good directional solidification.

141. Karl Singer,¹⁸ one of the exponents of superheated steel, is of the opinion that the general idea that hot steel has a greater tendency to crack arose from the fact that the hot melts, which are richer in oxide, usually are not treated any differently from the colder melts and that the deoxidation of the steel bath therefore is not carried far enough. The result of this is that the hot melts are more often poured with a higher oxygen content. The FeO, which is already separating out at the time the hot cracks are formed, lessens the slight tensile strength of the hot metal and thereby increases the danger of cracking. It is Singer's opinion that oxygen, or iron oxide, is a factor in formation of hot tears.

142. Regardless of the actual causes, the fact that steel poured hot produces satisfactory castings is established, and the reason for the greater tendency to hot-tear formation in castings made with hot steel cannot be attributed to the temperature. The pouring of a well-deoxidized steel from elevated temperatures, such as may be obtained from the electric furnace, will be found to be of advantage in casting thin-walled pieces with a tendency to crack.

143. If it is assumed that hot tears form during the transition of the steel from the plastic to the solid condition, then it can be shown that the temperature of the steel as poured is very important. For example, a high casting temperature will allow for a slow solidification of the casting, as the mold metal interface is superheated. Shrinkage stress will not arise suddenly, and a more uniform directional solidification can take place.

144. An advantage of low-temperature pouring is that a substantial elastic crust, capable of absorbing quite high tension stresses, forms while the mold is being filled. One of the disagreeable things arising from the use of low-temperature steel is the increasing proportion of entrapped gases. The thickly fluid steel should be limited to heavy sections. The principle of directional solidification should always be used.

145. Thin-walled sections that are liable to crack should be poured with superheated steel. Allowance must be made, of course, for the extra liquid contraction that takes place and for obtaining a well-made steel.

146. In summing up these points, it is the authors' opinion that the principal advantage of retarding solidification lies in the fact that an equalization of temperature takes place in the crust first solidified. By keeping the steel fluid in the mold for a longer time, directional solidification can be more readily controlled and suddenly applied stresses more easily eliminated.

(g) *Design.*

147. In a paper of this type the question of design should be given some attention. There is much to be said on so important a question, but to present a thorough study of it is not within the scope of this article.

148. There are as yet no definite rules of design which can be promulgated for every case, but it is possible to take specific cases and from them point out some generalities which might apply to similar designs. This has been done so frequently in the past that it is hardly necessary to mention more than a case or two until such a time as, we hope, more definite knowledge will be available to enable us to state what is the best design.

149. A study of proper design must include all of the factors mentioned previously in this article. It is scarcely to be expected that the most expert of designers will possess knowledge of the possibilities and limitations of foundry practice. Real cooperation should exist between the designer and foundryman in order that

both may profit by the economies in time and money thus effected.

150. Statements of this kind have appeared so frequently in recent technical papers that they are no longer novel. In spite of this fact, no great steps toward cooperation have appeared. It may be true in isolated cases that a genuine endeavor is made to bring about an understanding between the foundryman and his customer, but such cases are rare and do not apply to the industry as a whole.

151. In most cases the foundryman never sees the pattern until it arrives at his plant to be molded. Then, even though he knows it should be modified, he usually says nothing for fear he may lose the order or cast reflections upon his own ability. In many cases he would be money ahead if he did lose the order.

152. Most foundrymen will agree that a great many designs are unnecessarily difficult for foundry production. They are also agreed upon the fact that each casting is a problem in itself, and that foundry practice is an art and cannot be bound by definite rules or regulations.

153. The authors agree whole-heartedly with the first statement, but not entirely with the second. It is true that each casting as it is designed now is more or less a problem in itself. Even so, there are many general rules which apply to all castings and which every foundryman uses, subconsciously or otherwise.

154. However, there are at present no rules or limitations which apply to the designer. His main purpose is to design a casting that will perform the desired task with a minimum amount of weight and be at the same time pleasing to the eye. He has no knowledge of and as a rule is not concerned with the difficulties which may be met in production. He seldom realizes that in many cases he jeopardizes the entire casting for a small loss in weight or a slight gain in "beauty."

155. Accumulations of material are among the greatest dangers tending toward pipes and cracks. For constructive reasons it is not always possible to avoid them, but the designer can minimize the danger by providing gradual transitions from light to heavy sections.

156. Heuvers¹¹ has suggested a "simple and convenient method whereby the influence of individual cross-sections may be determined." This consists merely of inscribing circles within the cross-sections to be examined. The value of the method can readily be illustrated by examining the two radiographs (Fig. 13) of ribs of two sizes joined to sections of an identical cross-section.



FIG. 13—GAMMA RAY RADIOPHGRAPHS SHOWING INFLUENCE OF INDIVIDUAL ADJOINING SECTIONS.

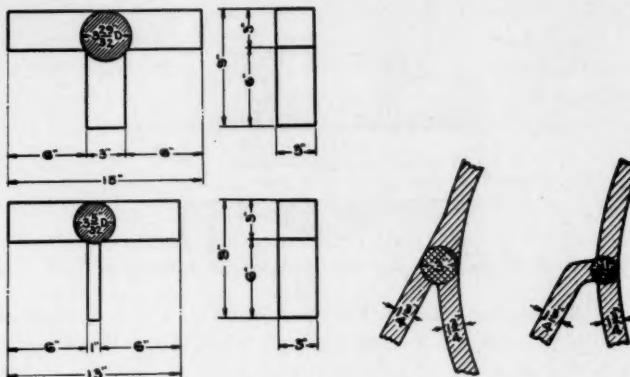


FIG. 14 (LEFT)—INSCRIBED CIRCLES AS APPLIED TO ADJOINING SECTIONS.

FIG. 15 (RIGHT)—APPLICATION OF INSCRIBED CIRCLES AS AN AID TO DESIGN IN A RIB SECTION. (AFTER HEUVERS.)

157. In the diagram (Fig. 14) these two sections are shown with the circles inscribed. It will be noted that the inscribed circles in each arm of the first T-section is 3 inches in diameter, whereas at the junction of the arms the diameter is $3\frac{29}{32}$ inches. In the second T-section the rib is but one inch thick and the inscribed circle has a diameter of $3\frac{5}{32}$ inches.

158. The radiograph of the larger cross section shows a flaw in the larger circle representing the accumulation of material. The second radiograph also shows a flaw, but one considerably smaller than the other.

159. Other instances where design could be improved are made clear by two further examples. The first, shown in Fig. 15, is merely a section of a tube designed with a rib or foot such as frequently serves to support pipe sections. The design usually chosen is represented on the left. A design recommended by Heuvers and used to some extent in Europe attains the same purpose, and although it is less pleasing to the eye, is decidedly preferable from the standpoint of foundry technology. The accumulation of material in the design on the left, as shown by the areas of the inscribed circles, is 80 per cent greater than that of the more correct design.

160. The second example (Fig. 16) is a portion of a commercially designed valve body. The right half of the cross-section shows the original design; the left half has been modified by the

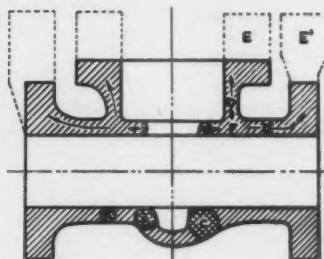


FIG. 16—DIRECTIONAL SOLIDIFICATION AS APPLIED TO A VALVE BODY.

authors. The dotted lines indicate portions that are to be removed by machining, and the arrows indicate what the authors believe would be the direction of solidification.

161. The changes in section on the modified side are all gradual and tend to reduce the accumulation of material or to produce controlled directional solidification. Using Heuvers' method of inscribed circles to examine the casting, it will be noted that circle *A*, representing the average wall thickness, has a diameter of $1\frac{1}{2}$ in.; and that circle *C*, representing the original design, has a diameter of $2\frac{1}{2}$ in., or an increase of 186 per cent in accumulated material.

162. By modifying the casting somewhat, a section as shown by circle *B* of $1\frac{3}{4}$ in. diameter may be obtained. This represents an increase in accumulated material of 36 per cent, which, although not perfect, is decidedly better than the original design.

163. The modifications on the flanges are similar to those pointed out by Wheeler²⁰ and, although they apparently cause an accumulation of material, simplify matters greatly by introducing controlled directional solidification. By examining the arrows on the original design it will be noted that, as solidification will begin at the thinner sections, points *D*, *D'* and *D''* will solidify first and solidification will proceed from these points. It then becomes apparent that, as far as feeding the hot point *F* is concerned, the heads *E* and *E'* are worthless. The head *E* cannot feed beyond the section *D*, the head *E'* beyond the section *D'*, and the appearance of a cavity at *F* is highly probable unless other precautionary measures are taken to obtain directional solidification.

164. If the modified design is examined it will be seen that solidification is more or less controlled and will proceed toward the heads and thus maintain an open avenue for "feed metal" until the head itself has solidified. To obtain this end the head

must be carried down the side of the flange, as otherwise the head itself might solidify before the heavy portion of the flange.

165. The studies on the formation of hot tears are in their infancy, and plans for a more complete research have been formulated. In the radiographic inspection of castings being purchased by the Navy, many conditions of external and internal hot tears have been observed. The fact that there are so many contradictory opinions regarding the nature and conditions of their formation has led the authors to express the above notes in the hopes that there will be greater cognizance of steel castings' most formidable defects, namely, the hot-tear formation cracks.

SUMMARY

166. Casting solidification involving the principle of equal cooling throughout is undesirable. Such conditions should be replaced by the more natural method of *directional solidification*. Controlled directional solidification is successful only when certain principles of casting design are adopted. These principles stress the importance of feeding smaller sections through heavier ones and by studying mass effect by the use of inscribed circles.

167. Metal contraction and hindered contraction resulting from mold resistance is responsible in the major part for the existence of hot-tear cracks in steel castings. Other important factors are cavities and the physical properties of the steel.

168. There are two types of hot-tear cracks, (1) the internal hot tear that develops during the solidification of the casting, and (2) the external hot tear that develops after solidification is complete but while the temperature is about 1300 degs. Cent. (2372 degs. Fahr.)

169. Intelligent application of certain principles in casting design is one of the most important of seven methods that may be adopted to obviate the danger of hot tears, or at least to limit the tendency of hot-tear formation.

ACKNOWLEDGMENT

170. The authors wish to take this opportunity of expressing their appreciation to Commander E. D. Almy, U. S. N., Director of the Naval Research Laboratory, for the interest and enthusiasm he has shown in the steel casting research project. It was largely through his efforts that a basic research on steel castings was inaugurated.

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WRITTEN DISCUSSION

R. A. BULL¹: The authors have chosen for their contribution a subject that always has been and will be of great practical significance. Fortunately for the needed progressive acquisition of definite knowledge of phenomena that occur in steel from the time it enters the mold until the casting reaches room temperature, the authors are adding scientific data to those previously ascertained on the same subject.

Accompanying the accumulation of any such purely scientific data, there should be an evaluation of it in the light of practical founding, to avoid unnecessary expenditure of time and money for prosecuting certain lines of investigation that might be carried on, and would have only what generally is termed academic significance. This comment is not made to discount the results of researches by the authors, but for the possible benefit of persons who may be prompted to investigate phenomena that occur in steel founding without fully taking into consideration the relationship that should be maintained between the development of facts of essentially scientific character and the economic application of such information.

I regret that I cannot supply useful experimental data accumulated in substantially the same way chosen by the authors. My compliance with their request to discuss the paper will take the form of comments prompted mainly by manufacturing experience. The following observations are made to strengthen the bond of co-operation between the more scientific investigators and the many steel founders who strive continuously to apply in practice the results of technical research.

For the purpose of preventing misunderstandings, the authors present several definitions. I submit a question regarding the term *pipe*, defined by the authors as a cavity in the solidified metal caused by contraction and formed during the change from the liquid to the solid state. The term as thus used is not employed universally in steel manufacture. Probably the phenomenon giving rise to the use of the word *pipe* to describe it in that industry, generally is thought of in connection with the contraction occurring in the top of an ingot. Presumably the word *pipe* was chosen for that condition because the nature of many of these contraction cavities in ingots resembles in some degree the hole in a pipe.

As the authors employ the term *pipe*, they mean, not a hole of partially cylindrical conformation, but the condition familiar to every steel foundryman, which often is rather hard to describe. More often than not in the steel foundry, such a condition is termed a *shrinkage cavity*. The word cavity can not be restricted correctly as to the general shape or dimensions of an opening. There are defective conditions in steel castings in which actual cavities are not visible to the naked eye but

¹ Consultant on Steel Castings, Chicago.

are uncovered by such a process as deep etching. That is one boundary of a wide range, the opposite one is the well defined, large opening. Purely in the interests of accurately descriptive nomenclature and to prevent misunderstandings by any who study such valuable contributions as the one by Messrs. Briggs and Gezelius, it is suggested that they substitute for the word pipe, the word *cavity*, or another word that will not connote a cylindrical hole.

With further reference to terminology, I would point out that the authors describe tears that occur after solidification "during cooling from solidification temperature to room temperature," as cold cracks. Such defects are not cold cracks, actually or in the sense meant by those words as customarily used by steel foundrymen. The employment of a term that is out of the ordinary and is less accurately descriptive may lead to misunderstandings. Every trade has its peculiar lingo, and the steel foundry has a phraseology of its own. It should be changed only for the purpose of clarification.

The authors appear to have found, after examining literature on casting solidification, that most men who have contributed information on the subject have finally pointed out that if a casting could maintain an equal cooling rate and solidify uniformly throughout, there would be less defects from cracks and that such a desirable condition may be reached by retarding the cooling of the thinner parts and chilling the thicker parts to equalize temperatures. The authors of the paper state that this "is a nice picture but is fundamentally unsound for several reasons." Then, they enter into the discussion of what they call directional solidification, and state that "solidification by uniform cooling must be discarded," and be displaced by directional solidification.

It seems desirable to comment on these references, because otherwise either a wrong impression might be created as to the actual opinions held by the authors or as to the actual facts.

It matters not so much by which one of several usable terms we describe the condition of progressive solidification and the fair approach to uniformity in temperature of various members in a newly-poured steel casting. It does matter considerably to have no one get the impression that it is unsound fundamentally to strive for a condition of fairly uniform cooling of sections, preferably always by intelligent designing, but capable of accomplishment to some extent by artificial means. Among these are the internal and external chills referred to by the authors, and in some cases other expedients, such as the flow of cold water through a portion of the mold. The latter effort is applied seldom and rarely would be advisable, or economical, or other grounds, but it has been made successfully.

A careful reading of the entire paper discloses no intention of the authors to discount the value of equalizing temperatures in the casting members, as the result of design or of chills. Indeed, the authors very properly emphasize these factors, but in the absence of these comments, some who may not study the entire paper carefully, might get the wrong impression. In any case, it seems to me that in the interests of truth,

the authors took the chance of being contradicted vigorously by declaring as fundamentally unsound the *reduction* of cracking defects by equalizing temperatures by design of molding expediments. Unquestionably, each of these procedures is fundamentally sound and distinctly helpful.

The authors state that it has been known for some time that certain processes produce a more fluid steel at lower temperatures, than do other processes. They refer particularly to converter steel, and point out that it is for this reason as much as any other that European foundries have maintained the converter practice. Following this declaration, the authors state that the degree of difference in fluidity of the best made open hearth steel and of electric steel, of acid steel and of basic steel, is not known definitely. They then ask a question as to whether temperature or fluidity is the more significant element, and say: "We are inclined to believe that fluidity is of greater importance, since the primary consideration of every foundryman is to run the casting completely."

There is no question in the minds of casting producers as to whether fluidity or temperature is more important. Necessities of production put fluidity firmly in the dominant position and no one worthy to be called a foundryman maintains that fluidity and temperature are always in absolute step with each other. It has been known for a long time that chemical composition is an important factor influencing the fluidity of molten steel. Thus, it is established that, other things being equal, the lower the carbon content in common steel and in most grades of simple alloy steel, the lower is the fluidity. Probably nobody knows all of the factors that influence fluidity beside the admittedly dominant one of temperature. Certainly, it appears that constituents other than the elements usually determined in chemical analysis have an important effect on fluidity. It has been the observation of numerous foundrymen that the regular grade of converter carbon steel for castings is more fluid at what appears to be a given temperature than is the regular grade of electric carbon steel for castings but it is most difficult to establish quantitative data on such points. The effect of any oxides might be counteracted or intensified by differences in chemical compositions as conventionally ascertained, or by analysis made to identify elements usually undetermined.

However this may be, I do not believe that the maintenance of converter steel foundry practice in Europe to much greater extent than in America at the present time is due largely to the superior fluidity of converter steel. From what information I have been able to obtain, the chief factor that has served to maintain the converter in some foreign steel foundries, notably those in England, has been the cost of electric power as compared with the cost of suitable materials for converter steel manufacture. Naturally, another element entering into all such matters, is the greater tendency of American manufacturers (up to 1930) to make investments of substantial amounts for installing the latest equipment. The more conservative attitude abroad in such respects is well known. Without some such comment as this, some might get an erroneous opinion regarding the claimed greater degree of fluidity of converter steel, over

steel made by any other process, for castings. And in any case, it may be said truthfully that superior steel, adequately fluid, for any purpose suitable for castings, may be made in electric furnaces; and that open hearth steel may be produced fluid enough to meet the demands of foundries imposing no more restrictions on themselves as to sizes of castings than are logical, as based on heat size and general shop equipment adapted for open hearth casting production. This is another way of saying that there are more limitations than fluidity, which place economical restrictions on open hearth plants as to the preferred sizes of castings to be produced profitably.

The authors have mentioned the discovery by an officer of the U. S. Navy recently that a fluidity test has been used in European steel foundries, particularly in France, to get an indication of the correct tapping or pouring conditions. The authors do not attempt to verify the claims made regarding the utility of such a fluidity test. A test casting for such a related purpose (applied to cast iron) was described before the A.F.A. in an exchange paper* contributed in 1927 by C. Curry on behalf of the Association Technique de Fonderie de France. This prompted experiments in some American steel foundries along similar lines. I know of several such plants which spent some time for this purpose, without reaching the conclusion that such a test casting offered advantages in connection with casting steel. It is conceivable readily that the device might be useful for cast iron, while not satisfactory for steel, because the shorter fluid life of steel as compared with iron naturally results in magnifying for steel, the importance of some factors which may be disregarded largely in connection with fluid cast iron. Reference to the European utilization of the scheme may make it worthwhile to mention briefly several factors that need to be maintained uniformly, such as the temperature of the mold when poured, the condition of the surface of the mold as to permeability and constituents that may produce porosity, and the condition of mechanically made outlets for risers or vents in molds. Another factor, admittedly difficult to regulate but important, is the method of filling the test mold with metal. If the French steel foundrymen have succeeded in taking care of all such conditions and have developed an economically usable test casting that guides them in getting the desired fluidity of their steel, they are entitled to much credit.

The authors have contributed some interesting information on the subject of shrinkage characteristics of acid and basic steel, after taking measurements under shop-laboratory conditions to develop usable information. It was found that the shrinkage characteristics of acid and basic open hearth steel were the same. My reading of the data covering this, makes me wonder whether the authors may have made the error of associating too closely the tendency to shrink and the resistance to shrinkage stress. These may be quite different characteristics. Probably it would be difficult to establish the difference between them, whatever it

* *Test Bars to Establish the Fluidity Qualities of Cast Iron*, Trans. A.F.A. vol. 35, pp. 289-306, 1927.

may be. I do not think we can assume, however, that a steel that has a high shrinkage characteristic always has solely for that reason, and in inverse proportion, a low resistance to cracking. The effects of sulphur and some other constituents may be very important. The foundryman knows that high sulphur makes steel red short, without noticeably influencing the degree of contraction or shrinkage of the steel. The foundryman also knows that high phosphorus increases fluidity, without apparently affecting the degree of shrinkage or producing red shortness, but developing cold shortness. It seems to me incorrect to apply the term "conducive to cracking" analogously with the shrinkage characteristic.

In speaking of contraction stresses the authors remark that it is "hard to realize that rammed sand can exert such a resistance through the contraction of steel, that fractures result." However, examination following pouring of many molds that have been oven-baked probably will eliminate any wonder that sand used for molding can produce hot fractures in steel castings. There are many molds formed of such materials, when baked, and when subsequently in contact with the molten metal, as can be appreciated readily as offering much more than the needed resistance to contraction stresses, to produce a "shrinkage crack." This has given rise to a tremendous range in the degrees of compressibility of steel casting molds.* The skilled foundrymen adapts the conditions to suit the design.

The authors make some statements that indicate their belief that the internal hot tear is not caused by mold resistance. However, it has been my observation that sometimes mold resistance causes an internal crack, if we include every crack that is invisible at the surface under that term. There are occasions when, seemingly because of the resistance of the mold, its obstructive influence on the plastic or semi-plastic casting at a critical point causes a rupture, after the skin of the casting has frozen so quickly as to resist fracture. In such cases, the hot tear extends to the very skin, and is in effect a surface tear, although the surface of the casting exhibits no imperfection.

The authors indirectly indicate their belief that an internal chill may be of such nature and the section of the casting may be of such size as to completely fuse the chill with the casting. I have yet to have demonstrated fully to me the complete fusion of any internal chill, when meticulously examined, as by deep etching. Such a scrutiny has disclosed the failure of very thin, clean chaplets to unite firmly and completely with very hot steel, in members of moderate thickness. Even so, unqualified condemnation of the internal chill is not justified. Such a device occasionally is very useful, as in forming a thick boss surrounded by thin members when the boss is to be drilled but needs to have thoroughly homogeneous metal lining the hole.

The authors rightly stress the need for cleanliness and dryness of all chills but I think they go too far in claiming that "whatever can be accomplished by chills may be attained by the use of external chills with less danger of harmful results." Certainly the use of internal chills is

not required for the satisfactory manufacture of the great majority of steel castings, and usually the external chill is the better one to employ.

The authors indicate a strong preference for steel castings poured hot, and are opposed to the practice which they attribute correctly to most experienced steel foundrymen, of pouring at as low a temperature as will successfully run the job. The authors declare that this procedure is "urgently in need of correction," but they present no evidence to support their opinion. Therefore, its value must be discounted in the face of the generally acquired belief, based on broad observation in making steel castings, that many difficulties from hot cracks are avoided without producing counteracting troubles of equivalent significance by pouring castings with metal merely hot enough to properly fill all mold cavities. I believe that long experience of any man in personally directing shop practice for producing large steel castings of such design as to make them susceptible to shrinkage cracks, would be apt to make such a person oppose the pouring of those castings with metal any hotter than required for the complete filling of the mold. Incidentally, this is one of those questions that are extremely difficult to settle by a conventional research procedure conducted for the purpose. Probably judgment formed after extended observations as to time and variety of castings affords the best practical guide for practice.

The paper terminates with very constructive comments on steel casting design. The authors should be commended highly for stressing this vital factor, which has never been given the attention it deserves, either by steel foundrymen as a class, or by consumers. Occasionally, some of us are inclined to criticise purchasers for requiring metallurgical monstrosities to be made in cast steel. However, we have no basis for criticising consumers on this count except when we have pointed out to them in a way that they can understand, those elements of design that must be observed when the objective is to secure a high degree of resistance in each member of the casting. Messrs. Briggs and Gezelius should be thanked by foundrymen for emphasizing such matters. They are to be commended and encouraged in connection with the careful investigation, now partially reported, which is apt to result progressively in valuable information for the steel casting industry and for the industries served by it.

F. A. MELMOTH²: Steel founders of the enlightened type, and there are indeed very many, have closely studied solidifying conditions, and, although as a class they may be somewhat diffident about entering into highly theoretical discussions involving complex physical proofs or explanations, they have tried quite honestly to apply their deductions to everyday practice. While steel founders are ready and anxious for the assistance and elucidation offered by purely theoretical deductions, the fact remains, and must be given due significance, that millions of tons of steel castings have been made of all sorts and conditions of types which have stood successfully all service demands made upon them.

² Detroit Steel Casting Co., Detroit.

The problem is definitely two-sided. The most accurate knowledge of fundamental facts regarding solidification will not make a good casting unless such casting is designed in the light of that fundamental knowledge. As a definite case in point, I would take the arguments raised by the authors on cooling conditions from the liquid to the solid state. The term used by the authors, *directional solidification*, is somewhat new in wordage, but the principle underlying it is well known, and is often referred to as *progressive solidification*. Viewed this way, our understanding would be that perfect design would permit of controlled solidification of a progressive type throughout the casting, the final contraction being cared for adequately by a reservoir of liquid metal, that is, a feeding head.

While sundry expedients permit of alleviation of the bad conditions set up by deviation from this main principle, not any one of them is free from jeopardy or unreliability of action. Therefore, we arrive at the point so strongly stressed by steel founders, that these matters become largely ones of educated design, and, in my opinion, this is the very heart of the whole matter, namely, the opportunity offered by design, for the application of sound, theoretical knowledge of solidification conditions.

Dealing specifically with one or two other points of the paper, in the first place, I would refer to the recommendation given for a standardized fluidity test. The drawbacks I see to this test are ones of manipulation. It is difficult to standardize time, and with a small sample spoon small variations are extremely effective. Speed of pouring is also a possible variable, although this might be mechanically overcome.

A reference also is made to the comparative tendencies to cracking exhibited by acid and basic, open-hearth steels. It does not appear essential to me to connect any possibly differing behavior in this respect with comparative shrinkage. Two heats of acid open-hearth steel made differently will manifest great differences in their behavior where cracking is concerned, and what often is referred to as *tender* steel is a well-known material. This suggests the existence of a variant tending towards intercrystalline weakness round about solidification temperature.

I am in agreement with the authors' formed opinion as to the temperature at which tearing occurs, and a practical confirmation exists in the fact that generally it is accepted that where relieving of any kind is practiced, such work must be done very early to be of any use. In fact, many old and experienced workers in the trade, for whose opinions I have the greatest respect, are confident that relieving is useless, as it cannot be done early enough to be effective.

The authors' remarks that mold resistance is conducive to hot tears could easily and truthfully read that mold and core resistance is always the cause of external hot tears where design is of a certain type. It is hard to imagine a casting tearing if it could solidify free from any interference with natural contraction; but we are obliged to have a mold. Not only so, but we must have one made of such material, and so firmly

constructed, that it will withstand the weight and erosion of very hot liquid steel.

Pipes are contributing factors to the formation of hot tears, but not necessarily directly and mechanically. A pipe is obviously a definite indication of the termination of some temperature gradient, and must be the high side of it. If this is so, then we must have a hot point during and immediately after solidification. The result is a stress concentration and the situation is aggravated by the presence of a weakening defect.

It is admitted that our knowledge of fundamental facts relative to the strength of steel at temperatures just below solidification is scanty. It is really more than that, it is almost non-existent, but deductions which are demonstrated by practice and which offer possibilities of effective remedial measures cannot be passed up as useless. For instance, perhaps we could justly describe the high temperature weakness to a lack of cohesion between newly formed crystals. If so, anything tending to lower this cohesion aggravates the weakness. Non-metallics of certain physical form, either in this way or by their action in effecting crystalline formation, may quite easily exert an influence. We all seem to arrive at the point, in any case, that pure clean steel tears less. Ductility of the individual crystals means nothing from the standpoint of hot tears, assuming that lack of cohesion between crystals has any significance. They cannot possibly stretch if they do not hold together. Perhaps there is some significance in the point that a steel made of 0.3 to 0.4 per cent carbon is more likely to be in decent condition than one of 0.15 to 0.20 per cent carbon, to which great care has not been given. A further point on this may also be interesting. Low cohesion, with large shrinkage spaces, appears more logical to me with very low carbon. Perhaps crystal shape and size have more significance than has been credited to them. Manganese also tends to crystal refinement in cast structures and it is a fact that in my own experience medium carbon, plus a generous manganese content, has been found best for certain intricate jobs of a type involving a high degree of mold and core resistance.

As I cannot imagine a condition where hot cracks form and refill from the main body of liquid steel, I do not find it easy to associate fluidity with a less tendency to hot tears, excepting, possibly, in the following fashion. A casting made from fluid steel is cleaner, smoother, and freer from surface imperfections, such as sand inclusions and surface enfoliation. It is realized that many of these types of defect can form points of local weakness, tending to stress concentration and rupture. In this way, fluid steel may contribute favorably to the hot tear question.

The only effect that I can see as a result of the use of preheated molds is a retardation of heat extraction, resulting in a general tendency for the casting to more nearly approach its solidification point as an entire mass before the formation of external envelopes of solidified metal; in other words, to reduce the time during which both liquid and solid phases exist together in the casting. This might certainly assist from the standpoint of internal tears. The characteristics of many core

materials prohibit a high degree of preheating, and I would not be very optimistic of any real advantages to be gained.

On the question of pouring from very high temperatures, there is a diversity of opinion among practical steel producers. Over a period of many years I have gradually, and entirely as the result of extended experience, formed the opinion that definitely cold casting is not good. Equally am I convinced that definitely hot casting is not good. The terms hot and cold are purely relative, and can be used correctly only in relation to the job in hand. Bearing this in mind, I would say, as a matter of personal opinion, that the correct temperature to pour any mold was that at which the steel completely and cleanly filled the mold cavity, giving a replica of the pattern in the desired detail. Anything over that is superfluous, means cost in steel making, danger in handling, excessive erosion of molds, increased absorption of mold gases, and a distinct tendency, in molds capable of being burnt hard, to an increased tearing hazard.

I would like finally to congratulate the authors on the thorough examination they are making of the whole subject. We cannot always hope to agree with them in full, but we can assure them of our earnest attention and co-operation in the work they are doing.

W. HARVEY PAYNE²: A study of metal solidification is always a most interesting subject to the foundryman. Fluidity of metal to be cast is very important and skilled melters accurately judge fluidity by the way the last metal flows from the test spoon. Temperature is judged by other methods, such as "freezover" tests and "color" for lower temperature range metals, or special measuring apparatus such as optical pyrometers, etc.

Fluidity of metals is a function of temperature *plus gas content*. A steel or iron heat that is fully degassed will have less fluidity for a given analysis and temperature. This is demonstrated time and time again. When an electric furnace replaces a converter in a steel foundry, we always look for the pouring crew to complain about fluidity of the better degassed electric furnace metal.

I have in mind two specific cases of fluidity of steels made under a highly reducing carbide slag in a basic furnace. The complaint in both cases was lack of fluidity. The quick answer in both cases was to give the heat a small *dose* of mill scale. I have seen this same thing in iron.

Easily over-killed steels in an electric furnace will have less fluidity than steels made in an open hearth simply because of the difference in gas content. However, steels of desirable fluidity may be made in either types of equipment by proper furnace operation.

Since a given casting design calls for certain fluidity of metal to run it, metal which is relatively too thoroughly degassed must have a higher temperature to flow the particular casting properly. Practice of many years tells us that it does not pay to pour the molds too hot.

Careful distinction should be made between the terms "de-oxidized"

² Pittsburgh Lectromelt Furnace Co., Chicago.

metals, "killed" metals and "degasified" metals. Metals may be deoxidized without being completely killed or degasified. The term *killed* metal usually refers to metal which does not *kickup* in the mold heads or risers. I feel safe in stating that killed metal does not necessarily mean thoroughly degasified or deoxidized, although it generally is accepted that they go together.

I wish to state emphatically that there is a great deal that we *do not* know about the effects of minor gases in metals, although we *do* know from real experience that gases cause some of our casting troubles, if we can always assign proper causes to conditions. I here refer to gases in metal which we pour as good metal.

A. W. LORENZ⁴: The authors are to be complimented on presenting a paper which so clearly outlines the mechanism of the formation of hot tears. This paper gives to the foundrymen, perhaps for the first time, an intimate understanding of the value of those preventive practices with which he has long been familiar.

I am inclined, however, to disagree with some of the authors' comments or hypotheses with respect to the effects of thoroughly-deoxidized steel. What is implied by the term "well-deoxidized steel," thoroughly-deoxidized steel? The following statements are quoted:

(1) "Steel poured hot will give excellent castings, and if it is a well-made, well-refined steel, *thoroughly deoxidized*, it will not have as great a tendency to the formation of hot tears as that which is poured cold. A great number of steel casting manufacturers do not make steel for castings but merely melt steel for casting."

(2) "The pouring of a *well-deoxidized* steel from elevated temperatures, such as may be obtained from the electric furnace, will be found to be of advantage in casting thin-walled pieces with a tendency to crack."

Beginning with the introduction of the electric furnace in the steel foundry, foundrymen made a conscientious effort to produce thoroughly deoxidized steel, but the results were disappointing. Instead of proving to be an ideal material, it was found to have a number of drawbacks, and it now is conceded generally that one of the chief of these drawbacks is a lack of fluidity. Electric steel men are well acquainted with the lifelessness and rapid setting qualities of highly-reduced steel, even when cast at temperatures considerably above normal. Consider on the other hand the high fluidity, as mentioned in the paper, of converter steel, the product of a distinctly oxidizing reaction.

These observations point in only one direction, namely, that fluidity is promoted by oxidation, and while the authors of the present paper imply that very little is known about the cause and control of fluidity, there is certainly a widely accepted opinion in the casting industry that fluidity and oxide content go hand in hand.

H. D. Phillips, in *Foundry*, March, 1933, states: "A steel lacking in iron oxide would be expected to have a higher fluid and solid contraction than an otherwise normal steel. This is so, and such steels cause more

⁴ Bucyrus-Erie Co., South Milwaukee, Wis.

rejections than normal steel due to pipe cavities, hot cracks, etc. In the absence of this iron oxide the requisite fluidity must be obtained at the expense of temperature."

Phillips further quotes Batty as having been first to call attention, in 1923, to the influence of iron oxide on fluidity.

During the past 2 years, the writer has had the opportunity of observing the performance of a basic open-hearth and a basic electric furnace working side by side on the same class of work. During this period, it was demonstrated beyond doubt that basic open-hearth steel was the more fluid, yet the basic electric steel was certainly "thoroughly deoxidized" under a carbide slag, while the basic open hearth process is rated as highly oxidizing. On the other hand, when the basic electric practice was modified to dispense with the carbide slag and finish up under slightly oxidizing conditions, there came, in the opinion of the men in the shop, a decided decrease in the number of cracked castings.

All of the above evidence is submitted to show that the authors' beliefs with respect to the necessity of obtaining thoroughly deoxidized steel are not compatible with their further conclusions regarding fluidity, unless by "thorough deoxidation" they mean merely that final deoxidation with silicon and manganese which is common to all steel making processes.

GEORGE BATTY⁵: A first reading of the paper impresses one with the magnitude of the subject which the authors have undertaken to investigate and it is certain that as investigation evokes facts, such facts may be found in conflict with opinions heretofore held by many of us in the steel castings industry. It is also well to remember that the interpretation put upon data may vary with individuals and so we come to have diversity of opinion in relation to facts.

In their opening paragraph the authors make a statement which will be disputed by some of us but which may be perfectly true so far as the authors themselves are concerned. I refer to the statement, "It is surprising to note how little experimental data are available on the actual physical processes which occur during the solidification and contraction of steel castings."

My belief is that a good many individuals are in possession of data, not much of which is published and available to all inquirers, which has led them to establish rules of procedure that promote the production of sound and serviceable steel castings of types wherein design is not an insurmountable barrier to the attainment of integral soundness of the cast structure.

For the purpose of discussing the paper it may be well for us to accept the definitions prescribed by the authors, but to do so with the reservation that such definitions are to apply only for the purpose of discussing the data presented in the paper. As warrant for this reservation, I would refer to definition of cold cracks. The authors state that, "Cold cracks are the result of large inherent contraction stresses pro-

⁵ Steel Castings Development Bureau, Philadelphia.

duced during cooling from the solidification temperature to room temperature. These cracks appear at comparatively low temperatures." This definition probably covers the source of origin of most cold cracks but other causes of cold cracking appear in actual practice.

I cannot agree that, to quote the authors, "From the information at hand it seems reasonable to assume that contraction takes place uniformly with the drop in temperature and that it (liquid contraction) is about 2 to 4 times greater than the solid contraction." Some years ago, I measured the contraction cavities in ingots which had been cast at a temperature of about 2800 degrees Fahrenheit and found that the cavities represented a volume approximately equal to 4.4 per cent of the cold ingot. At a later time, I made some investigation of the solid contraction of steel cast in sand molds and found that this varied from 2.2 to 2.4 per cent.

A reasonable interpretation of this linear contraction, therefore, is that the unrestricted solid contraction of a cast steel body is probably about 6.8 per cent as a maximum. If we compare this with the citation of the results of Benedick and Honda, we find that the sum of liquid contraction and solid contraction is not very widely divergent from the results of T. M. Service which indicates a volume change, from 2732 degrees Fahrenheit to ordinary earth temperatures, of 10.72 per cent.

It is not at all difficult to agree with the authors that the establishment of conditions which will permit a system of controlled directional solidification will promote soundness in steel castings. For years, I have been applying the principle of controlled, directional solidification to steel castings as far as the application of the principle was practicable within the limitations imposed by design. In fact, the first thing I did in a steel foundry in this country was to make a brake drum by a method which achieved its objective of abnormally high yield and unusual soundness entirely as a result of controlled directional solidification.

The authors are certainly correct in stating, "Directional solidification is a process that is intimately involved with the design of the casting." They might even go further and state that insofar as design impairs the practicability of achieving directional solidification, it connotes imperfection.

The tapered ingot, small end down, is the most popular exposition of controlled directional solidification. This principle of ingot production was, I believe, first applied by Harry Brearley and has been propagated worthily in this country by Gathman. Without entering upon any discussion of the relative merits of properly proportioned taper in steel ingots, it is sufficient to say that they largely have superseded parallel sided ingots. If a steel founder gets anything to cast as simple in form as is a parallel sided ingot, he believes he has a very simple proposition before him as compared with the complexity of the average steel casting. This statement alone is sufficient justification for the authors' contention that directional solidification is a process that is intimately involved with the design of the casting. The fact that many steel castings necessarily must be of a complex nature, does not affect the rectitude of the author's asser-

tion. It also supports their contention that any conception of a mold rapidly filled with metal of a low temperature to produce uniform freezing is impracticable fundamentally.

Brearley expressed the solidification of an ingot as a progressively thickening solid envelope and illustrated thereby the advantages of the tapered ingot cast small end down. With a complex steel casting, as the authors point out, we have to invoke the aid of either feed heads on isolated heavy sections or the application of chills to heavy sections when these cannot be fed practicably from risers. The practical foundryman knows all this and will welcome any assistance that can be given to him. If, at some future time, the system of chilling inaccessible, relatively heavy, masses imposed upon a generally light structure is reduced to an exact science, the practical foundrymen will welcome the scientist. At the present time, the practical man declines to be abashed when he is told that he has not done a thing which no one, as yet, has proved practicable. I have good reason to believe that many operative foundrymen are now achieving results, from the careful application of external chills, that make available serviceable castings of a design that is metallurgically appalling. That such castings are merely commercially sound and not metallurgically sound is no real indictment of the steel founder. It merely confirms the authors' statements in reference to the significance of design.

The authors make a statement which would appear to me to demand some modification. I suggest that they would be correct if they stated "In other larger sections where the concentration of mass is much greater, no skin at all will be formed near the ingate or ingates during the pouring interval."

In our consideration of the principle of controlled directional solidification, we must remember that the mold has some significant influence. Most of us have observed the sluggishness of metal in risers or feed heads remote from the gate of a casting. This metal has been cooled by contact with the mold and we must not assume that what we see in the risers is the first metal that entered the mold. In traversing the mold, some of the metal becomes so cool and sluggish that it adheres to the mold with which it is in contact and so produces a solidified envelope through which metal flows to fill parts of the mold still more remote from the ingate. From this fact certain inferences present themselves.

1. That the heat lost by the metal which fills the parts of the mold remote from the ingate has been transferred largely to the mold.
2. That a temperature gradient is produced in the mold as well as in the metal.
3. That in certain types of castings, gated in an orthodox manner, this temperature gradient in both mold and metal is inimical to the proper feeding of the casting.
4. That in the region of the ingate or ingates, the mold will be heated abnormally, will tend in that locality to insulate the casting over a period of time, and will promote a relation of mold to casting that may be prone to cause some local unsoundness.

The system of controlled directional solidification, therefore, must be

concerned with gating procedure as well as with design, chilling, and heading.

Controlled directional solidification is easy to achieve on some types of castings and produces a soundness that was somewhat difficult and expensive to obtain in terms of orthodox practice but we must recognize that there is a limit to what can be done, in producing castings of complex design, in the way of promoting controlled directional solidification.

The experiment performed by casting a 6-inch diameter sphere, and a parallelepiped of identical volume but of 50 per cent greater surface area, is interesting in showing that the rate of thickening of solidified envelope is practically the same in each case. This is not surprising when the two castings are visualized. They are both relatively massive and the practical foundrymen would not expect any great difference in the rate of shell formation except a little thickening at the angles of the parallelepiped. The significant point of this experiment, to the steel foundrymen, is that in each period of elapsed time the parallelepiped mold promoted the solidification of approximately 50 per cent more metal than did the spherical mold.

What the practical foundryman has been concerned about when he has referred to the fluidity of the steel, really has been the *castability* or *runability* of the metal. Particularly has this been the case in the production of light steel castings where, because of light section, the chill action of the mold is most quickly and powerfully exerted upon the metal to promote cooling and the concomitant sluggishness, and where it is unusual to encounter designs which permit the forcing of relatively sluggish metal into the remoter parts of the mold by the imposition of ferrostatic pressure in the downgate. This *runability test* is, in foundries producing light steel castings, probably best considered as a *shankability* test, the ability of the steel to stand exposure to the atmosphere in handshanks or small pouring ladles, and in this respect the *melter's count* test frequently provides a distinctly satisfactory indication of the ability of the metal to fill the molds properly.

The authors, however, now postulate a type of fluidity which may be distinct from sheer runability. I do not attempt to dispute such a postulate but would say that steel of the type referred to as *over-reduced* seems generally to give more trouble from secondary pipes and externally visible hot tears than does steel of similar composition, as revealed by conventional analysis, which has not been over-reduced. I have been inclined to believe that this was due to a higher freezing point, a more restricted fluid range, of the over-reduced steel and, reverting to one of my earlier comments, would say that the steel which showed the highest linear contraction was over-reduced considerably.

In reference to the horizontal spiral fluidity test mold, I would suggest that to make this a more definite indication of runability, some variables, which are at present obvious in its application, be removed. It would appear advisable to take a sample of metal much greater in volume than will completely fill the mold and to produce, at the head of the downgate,

a reservoir to hold this metal. The downgate should be plugged by using a rod of baked core sand, as a stopper is used in a bottom pour ladle, and the reservoir filled with a sample of steel taken from the furnace. A definite time interval should be established from the moment the spoon leaves the furnace until the stopper is lifted to allow the metal to run into the mold. Even this procedure is not immune entirely from variables, but it appears to me that the advocated procedure would tend to reduce the magnitude of the variables and would thereby permit the system to provide a more accurate indication of the runability of the metal.

On the subject of contraction in its relation to the temperature of hot tear formation, the authors' citations from the work of Korber and Schitzkowski will appeal to many practical foundrymen, but in order that this information may be useful to foundrymen they must be impelled to provide the means for the substantial elimination of mold resistance to the contracting casting. The authors have quoted a number of the means that are used by practical foundrymen and have proved thereby that "by rule of thumb or as experience has dictated" the practical man has determined the cause and prescribed a substantial cure.

The temperature measurements on the metal of a 6-inch diameter cast sphere along with temperature measurements of the sand, considered with the results accruing from the experiments on the rate of skin formation with sphere and parallelepiped, must be of considerable interest to practical foundrymen in their consideration of the principle of controlled directional solidification. I would express the hope that the authors will repeat the experiment with the 6-inch diameter, cast sphere and determine sand, or mold, temperatures in the wide open space which at present appears on Fig. 6 between the location of the thermocouple at $\frac{1}{8}$ inch from the mold-metal interface and the thermocouple at 1 inch from the mold-metal interface. The relation of maximum temperature of metal in mold to minimum temperature of metal in contact with mold, in the 6 inch sphere casting, is of great interest in considering the origin of hot tears. However, we must recognize that in a more complex casting we may have greater differences than those indicated by the two upper curves in Fig. 6.

By introducing the matter of internal hot tears radiating from enclosed contraction cavities, the authors provide a distinctly contentious subject for discussion. I have no doubt that a good deal of seemingly contradictory evidence will be offered both in relation to metal composition, pouring temperature, and the feasibility of attaining controlled directional solidification. I have a case in mind where some castings of a rather difficult type were produced satisfactorily, under operating conditions which substantially achieved controlled directional solidification, when the metal was of the medium manganese type and contained about 0.35 per cent carbon. These castings weighed about 7000 lbs. each and when a repeat order came along it carried with it a prescription of composition limiting carbon content to 0.22 per cent and manganese content to 0.65 per cent. The molds and cores were made in precisely the same manner as had applied to the production of the first lot of castings, and the gating and heading system, prescribing controlled directional solidification, also was

identical, but it was found, upon machining the second lot of castings, that they evidenced tears of the type indicated in radiographs 9 and 10. These tears were presumed, from their location in the casting, to be the terminal vestiges of a secondary pipe. The foregoing is submitted as evidence in favor of what foundrymen consider a more inherently fluid material than ordinary plain low carbon steel.

I have had several other experiences of a similar or related type which tend to confirm an assumption that the more inherently fluid steels are less prone to internal hot tears and are mutable more readily to the application of the principle of controlled directional solidification. At the same time, I would say that controlled directional solidification is achievable more completely with hot metal than with cool metal and I have largely committed myself to the practice of pouring "as hot as is practicable" rather than as cool as possible. You will note that I do not say "pouring as hot as possible". The word "practicable" takes cognizance of the ability of the molding materials to withstand the thermal and physical impact of the cast metal.

The relation of a small external tear to an internal contraction cavity, particularly at the radiused junction of a flange with a body, is something with which most practical foundrymen are acquainted. I well remember that in the early days of my association with the steel castings industry, Melmoth pointed out to me a slight tear at the root of a flange and stated that it indicated not only an internal defect but also evidenced what he ineuphoniously described as "lousy gating and heading."

In considering the formation of external cracks or hot tears of the ragged-edged type, we should, particularly in castings of complex design, consider the influence of the dilatation phenomena. Of course, it is quite easy to say that the remedy for external hot tears lies in providing adequate mold relieving but it is not practicable easily in all cases to ensure that the relieving provisions become operative at a sufficiently early time. In some cases, it is possible to achieve the desired result by a modification of design and thereby eliminate the necessity for relatively laborious procedure in providing for mechanical relieving of the mold.

The particular instance I have in mind in making such an assertion is related to the production of truck wheels having spokes of a cruciform section. Quite a number of different wheels were designed with spokes of a Y-type, cruciform in section, the upper members of the Y being lighter than the lower member of the Y. These wheels all were gated at the hub and a deplorable proportion of the castings cracked either at the junction of the spoke with the hub or closely adjacent to the junction of the two upper members with the lower member of the Y. At each of these two points there was a concentration of mass and obviously these were the proper points at which to expect rupture.

An absolute cure for this cold cracking trouble of the spokes was found to lie in curving the upper members of the "Y" so that these could, by straightening slightly, absorb the stresses imposed on the spokes by the relatively late contraction of the hub and inner ends of the spokes as

compared with the previously achieved major contraction of the rim of the wheel. The molds in which these wheel castings were produced were, necessarily, green sand but they received no mechanical relieving.

Another extremely contentious subject is introduced by the authors when they use the word "chills". I have come to the conclusion that it is desirable, wherever possible, to avoid the use of internal or "cast-in" chills and substitute external chills to contribute to that most desirable thing, controlled directional solidification. The cast-in chill over which any considerable volume of metal flows is likely to be non-effective. The cast-in chill is most effective when metal merely flows to it, surrounds it, and very little metal flows over or past it.

Where I have been unable to escape, from reasons of economy, the use of an internal or cast-in chill, I have come to adopt the principle of having twice as much of the chill embedded in the mold as protrudes into the mold cavity to become part of the casting. This provision increases the efficiency of the chill as a true medium of promoting soundness.

External chills have, to my mind, distinct advantages in comparison with internal chills but they are a much greater expense and are more difficult to apply.

There are a good many other points in the paper which call for discussion and I have no doubt that many foundrymen will wish to go to the mat with the authors who have, in their paper, said certain things in derogation of some steel foundry practices. The authors should not be alarmed or perturbed if their views evoke fervid criticism and again I would offer the suggestion that most of our differences may be related to our translation or interpretation of the evidence which comes before us.

The most proper tribute I can pay to the authors is to say that although their paper may be read in less than an hour, it cannot adequately be discussed in ten hours.

REPLY BY AUTHORS TO WRITTEN DISCUSSION

C. W. BRIGGS AND R. A. GEZELIUS: We appreciate Mr. Bull's commendable discussion as his long association with the industry has given him many practical observations. We accept his term of cavity in preference to the word pipe as it expresses more accurately the thought that we wish to convey.

We note with surprise that he has stated that it is an established fact, other things being equal, the lower the carbon content in common steels the lower is the fluidity. We knew that this assumption was ever prevalent. However, we were not aware that it was an established fact, especially considering how nearly an impossibility it is to get other things equal. Experimental data on this subject would be welcomed by us.

There is no doubt Mr. Bull is right concerning the economical aspects being important considerations in the maintenance of the converter prac-

tice in Europe. We merely were expressing an opinion that had been obtained by Captain Shane from certain foundry operators in Europe that it was important when related to a study of fluidity.

Along with his discussion on fluidity, Mr. Bull pointed out the need for maintaining a uniformity in mold condition variables. We believe that if care is exercised this can be done. At least, we are able regularly to produce a uniformity of conditions. It is our experience that the greatest difficulty is the method of filling the test mold with metal.

We agree with Mr. Bull that the tendency of steel to shrink and the resistance to shrinkage stress might be quite different characteristics. There are no data available now that would allow for a differentiation of the two. However, we hope to present such data in the future.

We are unable to accept Mr. Bull's statement that mold resistance sometimes causes an internal hot tear. In the first place, mold resistance acts to form hot tears only when contraction has begun. Contraction would be greater in the skin of the casting and it would be at this place that the stresses first would centralize. If the stresses were sufficiently great to overcome the strength of the metal, the crack from the surface inward would form. From the available information, it does not seem best to assume that the solidified skin is capable of plastic deformation thereby transferring the stress centralization to the interior less strong metal, since microscopic examination of tear formations have shown that the cracks exist around the grains.

Secondly, we were wondering how Mr. Bull, from his observations, could correlate definitely the internal crack with mold resistance especially when the lack of directional solidification is so ever prevalent in a casting.

We are exceptionally well pleased with Mr. Batty's discussion for he has brought out many points that are valuable in a presentation of a subject of this type. We agree with him when he suggests that the definitions used may not be entirely accurate from the standpoint of foundry phraseology and we thank him for pointing out for us that they apply only for the purpose of discussing the data presented in the paper.

In discussing the rate of skin formation Mr. Batty's suggestion is correct. However, we have found in our studies of 9-inch spheres that in some cases no skin at all has been formed when the sphere was emptied immediately after it had been filled. In some cases a lacelike skeleton remained.

In referring to the remark that liquid contraction is about 2 to 4 times greater than the solid contraction and presenting a dissenting opinion, we believe that Mr. Batty misconstrued the text and confused liquid contraction with freezing contraction. Of course, the total contraction is not 2 to 4 times greater than the solid contraction. This is largely an error on our part as the phrase "per degree drop in temperature" was omitted from the copy presented to Mr. Batty. However, the data he presents is in agreement with that of Benedick and Honda with respect to the contrac-

tion occurring while the steel is passing from the liquid to the solid state.

The statement to which Mr. Batty refers does not concern this contraction, but that of the liquid steel cooling from the pouring temperature to the temperature at which solidification begins. It is evident that Mr. Batty has divided solidification into two classes; fluid contraction and solid contraction, while we have divided it into three classifications of liquid, freezing and solid contraction. It is in this misunderstanding that the confusion lies.

In regard to the comments by Mr. Lorenz, Mr. Bull, Mr. Melmoth, and others on the subject of pouring hot we have the following further to say. Our suggestion concerning the use of superheated steel arises from the fact that we firmly believe in applying the principles of directional solidification to its fullest extent. Again, we point out that these principles involve the use of metal as hot as is practicable, pouring as slowly as is practicable to avoid the entrapping of solidified or oxidized films, and in using the excess temperature of the metal to produce a pronounced temperature gradient in the mold as well as in the casting itself. We have not taken the stand, as some of those who have discussed this point wish to infer, that all castings should be poured with superheated steel. In fact, reference was made only to thin walled pieces. The value of this section may be discounted as Mr. Bull suggested as we have offered no experimental data other than sound reasoning to support our contention but may we also point out that no one else during this discussion or in other treatises on this subject has offered experimental data to establish the acceptable "pour her cold" slogan. Perhaps if some of the experienced steel foundrymen had extended their observations with gamma ray radiography they would find some interesting data in regards to hot tear formation. A great number of failures of hot poured pieces have been attributed to the tendency of such pieces to expose a fine secondary pipe, or contraction cavities, immediately below the heads, but what is not so well known is that a cold poured cavity has similar, or worse, cavities or series of hot tears that are more deeply located in the casting. These are entirely missed by surface observation.

In reply to Mr. A. W. Lorenz' discussion, we used the term of thorough deoxidation to refer merely to final deoxidation with silicon, manganese, or other deoxidizers. We were not by any manner or means referring to the use of over-reduced steel as we are certain such steel would not be considered as a well made, well refined steel. Nor are we definitely confirmed to the fact that a well refined steel is one that is completely deoxidized under a carbide slag.

We appreciate Mr. Batty's statements on the use of high pouring temperatures. It is encouraging to note that he considers the use of superheated steel as practicable in certain cases.

We quite appreciate Mr. Melmoth's point of view when he contends that "millions of tons of steel castings have been made that have successfully stood all service demands made on them". This is true and yet no one can deny that "wasters" are still found at times, in even the best of foundries. It is to eliminate, if possible, these "wasters" that this re-

search program was inaugurated. These defective castings are the important factor in this research. We are interested primarily in the most prevalent defects, their causes and their remedies. If, therefore, in our discussions we dwell mostly on defects and defective castings it is not because we have lost sight of those that "have stood all the service demands placed on them" but rather that we have, for the moment, placed them in the background.

We do not believe that because we have emphasized defective castings that Mr. Melmoth, or any other foundryman, should feel that it is any slur upon the industry or that we have not "given due significance" to the fact that good steel castings have been and can be made.

We do not claim that the term "directional solidification" is new in wordage. If Mr. Melmoth prefers "progressive solidification", we see no objection to its use. Our search of the literature showed us that the term "directional solidification" was more commonly used. We are making no claims for establishing any underlying principle; we are merely asking for its universal application.

We believe that Mr. Melmoth misread the text of the paper when he refers to the fact that a recommendation was given for a standardized fluidity test. No such recommendation was made. In fact, to quote the text the following points were emphasized: "The use of a fluidity test instead of a stop-watch test seems to be the ideal way of obtaining advance knowledge of the mold-running abilities. However, at the present time it appears that a definite correlation of conditions will be difficult to obtain."

We value Mr. Melmoth's opinions relating to cracking and the strength of steel at temperatures just below its solidification.

We are sorry that Mr. Melmoth is unable to appreciate the work of Singer and Bennek in regards to the filling of tears with fluid metal as it appeared to be carefully executed and their conclusions did not seem to be out of line with their findings. We feel that the case is rather the expectation than a regular procedure, but we can imagine the condition.

Mr. Payne states emphatically that fluidity of metals is a function of temperature and gas content. With this we are not prepared to agree. Time and further experiment may prove Mr. Payne's contention to be correct but at this time we can find no data that substantiates this claim. C. M. Seager, Jr., and A. I. Kryniitsky* point out conclusively that, in all the metals they studied, fluidity is a function of temperature but no mention is made of gas content. Nor can any conclusions be drawn from this work that definitely prove the effect of gases on the fluidity of metals.

We are at a loss to understand Mr. Payne's differentiation between "killed," "degasified," and "deoxidized" steel and, therefore, have no comments on this portion of his discussion.

* Reference on bibliography, p. 424.

ORAL DISCUSSION

H. R. ISENBURGER^a: About two years ago, I presented a paper^b before the American Society of Mechanical Engineers where I showed x-ray pictures illustrating hot tears. We distinguished between hot tears and shrinkage cracks because there is a difference between the two defects.

A shrinkage crack may not be as serious as a hot tear. A hot tear is a very bad defect in a casting. The illustrations contained in the A.S.M.E. paper mentioned previously demonstrate this point. A hot tear has a much sharper outline of an image in an x-ray negative than a shrinkage crack.

F. A. MELMOTH: A report^c has been published recently by a group of scientists in England on steel castings. One of the conclusions that they reached is that quick pouring is necessary for sound castings. However, Mr. Briggs mentioned using super-heated steel and pouring slowly. In regard to the question of high temperature and solidification, we formerly called it progressive solidification. If you desire to produce a temperature gradient, it does not matter whether the gradient is on the low side of the temperature or on the high side, where it can cause a lot of incidental troubles as long as it exists.

GEORGE BATTY: I feel that it is important to create as great a temperature differential as possible in both mold and metal.

F. A. MELMOTH: You may start with 3000 degrees Fahr. and get a temperature grading from 3000 to 2500 degrees Fahr. or start with 2800 degrees Fahr. and get a similar grading but at a lower scale.

G. BATTY: The degree is not the same because of the effect of the metal on the mold. That fact has been established substantially.

C. E. SIMS^d: In discussing the matter of hot cracks, the statement has been made several times that hot cracks are produced after solidification. I wish to suggest the probability that they form before complete solidification. I believe that the hot cracks are formed while there is still a liquid eutectic in the steel. The most characteristic feature of hot cracks is the fact that there is no crystal deformation. The appearance of the fracture shows undeformed crystals similar to those found in a shrinkage cavity. This shows that there was practically no resistance to the metal tearing apart. Only a liquid would have such low strength. It is entirely conceivable that some of the eutectics in steel may be liquid as low as 2300 degrees Fahr. We are familiar with the phenomenon known as burning of steel in forging which is due to the fact that the steel is heated to the point where the eutectic becomes liquid.

The cure for hot cracks may lie in controlling these eutectics. There seems to be a conflict in the statements that certain alloys and high carbon decrease the tendency for hot cracks. It would seem that these might increase the tendency for eutectics, but there is the other chance that they

^a First Report of Steel Castings Research Committee, British Iron and Steel Institute (Special Report No. 3, 1933).

^b St. John X-Ray Service Corp., New York City.

^c Radiographic Inspection of Metals, Mechanical Engineering, vol. 53, pp. 729-735, 1931.

^d American Steel Foundries, East Chicago, Ind.

may lengthen the freezing period of the steel as a whole and bring general solidification nearer to the freezing point of the eutectics.

Deoxidation of steel might be another factor. The words perfect deoxidation or perfectly deoxidized steel too often have been used loosely. It is probable that perfectly deoxidized steel is not thoroughly deoxidized steel. It may be steel that contains considerable oxygen. There is some indication that the quantity of oxygen retained may have some control on these eutectics and indirectly have an influence on hot tear formation.

I agree that it is desirable to attempt to distinguish between hot tears and shrinkage cracks. It is my opinion that hot tears are caused by shrinkage. That is especially true in corners. There can be no method by which the particular stress which opens up the steel is produced than shrinkage. Both are due to stresses caused by non-uniform contraction of the steel.

REPLY TO ORAL DISCUSSION BY AUTHORS

C. W. BRIGGS AND R. A. GEZELIUS: In reply to the oral discussion, we wish to point out to Mr. Isenburger that there is no difference between a hot tear and a shrinkage crack in form or appearance. The term is used synonymously. We note that Mr. Isenburger claims to be able to distinguish between them even though he is unable to define and explain the characteristics of each.

We appreciate Mr. Sims' comments on the formation of hot tears. His views are similar to those expressed by Korber and Schitzkowski. It is true that the deoxidation of the steel may be an important factor in the formation of hot tears.

Mr. Melmoth mentioned the report of the Steel Castings Research Committee,* of the British Iron and Steel Institute as confirmation of his opinion that quick pouring is necessary for sound steel casting.

After a careful study of this report, we fail to see how any definite conclusion as to the effect of pouring speed on cracking can be drawn from the work done. The objections that we have to using this work as a basis for a definite opinion are: (1) the several castings were poured at different foundries under different conditions; (2) no mention is made as to sand conditions or molding procedure in any portion of the report and: (3) the table upon which the conclusion as to the effect of pouring speed was based can be amended so that other causes might just as logically explain the cracking in the different castings.

It might be well to take up these points in more detail.

(1) There is no logical reason to assume that the steel in each case

* *First Report of Steel Castings Research Committee, British Iron and Steel Institute (Special Report No. 3, 1933).*

presented the same resistance to cracking. The heats from which the castings were poured were not only made at different foundries but the composition, according to the analysis given on page 35 of the report, varied as follows: C., 0.17-0.27 per cent; Si., 0.10-0.41 per cent; Mn., 0.59-1.10 per cent; S., 0.002-0.028 per cent; P., 0.008-0.034 per cent; Ni., 0.05-0.56 per cent and Cr., 0.02-0.13 per cent. The castings were gated and fed at different points of the castings and the sizes and shapes of the heads varied considerably. This difference in gating and feeding was discounted in formulating the opinion on the effect of pouring on cracking.

(2) As nothing is stated concerning the sand conditions or molding technique, the reader has no way of knowing whether or not an attempt was made to have these conditions the same in each case, that is, whether all of the molds were hand rammed, machine rammed or both, whether relieving of any kind was used in any, all, or none of the molds, or whether the collapsibility of the cores was comparable.

(3) These objections become more important when the tables upon which the conclusion was based are studied. We have reproduced the table which appears on page 44 of the report as Table 2 and have taken the liberty of adding to it two other columns of data contained elsewhere. These added columns show the identification letter of the firm making the casting and the type of steel used.

If only the first three columns are studied, it would appear that the speed of pouring was the vital factor as far as cracks in this particular casting were concerned. If the last two columns are taken into consideration, one notes that castings 8 and 9 were not only poured faster but at the same foundry out of the same heat of steel. Castings 1, 2, and 3,

Table 2
DEFECTS PRODUCED BY VARIOUS POURING RATES*

Casting No.	Time to fill mold secs.	Original Table		Added Data	
		Cracks	Maker	Type of Steel	
8	14	{ None	{	N	Basic Electric
9	14			N	Same cast as 8
2	25	{ Slight, at ends of	{	F	Basic Electric
1	27	{ 3 bulkhead brackets	{	F	Same cast as 2
3	28		{	F	Same cast as 2
16	30	{ Slight, at ends of	{	M	Basic electric
15	37	{ Slight longitudinal crack in bore	{	L	Basic electric
5	39	{	{	G	Acid. O.H.
13	40		{	L	Basic electric
4	47	{ Serious	{	G	Basic electric
14	50		{	L	Same cast as 13

* First Report of Steel Castings Research Committee, British Iron and Steel Institute (Special Report No. 3), p. 44.

which were cracked only slightly, were all made at another foundry and cast from the same heat. Considering the four castings that were seriously cracked, one finds that castings 4 and 5 were made at one foundry and that castings 13 and 14 were poured from the same heat at still another foundry.

If all of these data are considered, it is rather difficult to state that pouring speed alone was responsible for the cracks. That might be the case, but as the seriousness of the defects seems to depend also upon the particular firm that made the casting, it is just as logical to assume that the sand conditions and molding technique at the several foundries were the vital factors. As the seriousness of the crack also is associated with a particular heat of steel in almost every case, the steel itself may be the important consideration.

Properties of Gray Cast Iron as Affected by Casting Conditions*

By C. M. SAEGER, JR.† and E. J. ASH, ** WASHINGTON, D. C.

Abstract

A preliminary report of an investigation to determine the effect of maximum heating temperatures on the physical properties of different types of cast iron is given in this paper. It describes the methods and results obtained in a study of three types of cast iron. Four heats, all melted in a high-frequency induction furnace, were made for each class of iron with maximum heating temperatures of 1400, 1500, 1600 and 1700 degs. Cent. (2550, 2730, 2910 and 3090 degs. Fahr., respectively.) Four duplicate transverse test bars of different diameters were cast from each heat in a dry-sand mold and the following properties determined: transverse breaking load, deflection, modulus of rupture and modulus of elasticity, hardness, density and microstructure. The shrinkage and the running qualities of the irons were also investigated. The transverse breaking strength for each iron changed with maximum temperature to which the liquid iron had been heated and, for two types of iron, the maximum strength for each size of bar was obtained at a different maximum temperature of the liquid metal. In general, the density of the solid metal and the linear contraction increased with heating temperature of the liquid metal, whereas the density in the liquid state was not affected. The running quality of the irons investigated was apparently not affected by the maximum heating temperature, but was found to be a function of the liquidus temperature, being inversely proportional to the solidification range. The microstructure of the 1.2-inch bars indicated that irons of highest strength had relatively small graphite flakes and a pearlitic-sorbitic matrix.

INTRODUCTION

1. The efforts of foundrymen to improve cast iron date back to about 1860, when Joseph S. Seaman of Pittsburgh discovered

* Publication approved by the Director of the Bureau of Standards, U. S. Department of Commerce.

† Physicist, U. S. Bureau of Standards.

** Assistant Scientist, U. S. Bureau of Standards.

Note: This paper was presented at one of the sessions on Cast Iron at the 1933 Convention of the American Foundrymen's Association.

that the addition of steel scrap to an air-furnace charge resulted in a decided improvement in the strength of the iron.¹ In 1906, Goerens² pointed out that gray cast iron is similar in structure to a steel containing embedded graphite, and that the best physical properties should be obtained when the matrix is pearlitic in structure.

2. Numerous investigators have shown since that the strength of cast iron is a function of the amount and distribution of graphite, the optimum properties being obtained with a pearlitic or sorbitic ground mass in which the graphite is disseminated as small globular particles. Improvement in the strength of cast iron has been brought about by control of composition, of melting and foundry practice, and by heat treatment of the solid castings.

3. This paper is a preliminary report on a study of the effect on the physical properties of several different types of iron, of heating the iron to various temperatures above the melting range.

MATERIALS AND METHODS OF CASTING TEST SPECIMENS

1—Materials.

4. The following types of iron were used: *A*, a typical stove-plate iron; *B*, a mixture of *A* and 20 per cent of commercial open-hearth ingot iron; *C*, a high-carbon, low-silicon, low-phosphorus pig iron, often referred to as a heat-resisting gray cast iron. The compositions of the pig irons were as follows:

	Total Carbon, per cent.	Graph. Carbon, per cent.	Comb. Carbon, per cent.	Manganese, per cent.	Phosphorus, per cent.	Sulphur, per cent.	Silicon, per cent.
<i>A</i>	3.55	3.15	0.40	0.24	0.82	0.021	2.73
<i>C</i>	3.79	3.13	0.66	0.73	0.12	0.06	1.32

5. *Molding.* Eight cylindrical transverse test bars of four sizes were molded in a 3-part cylindrical flask with the cheek extending the full length of the bars. Lumberton sand (A.F.A. grade 3-G³), tempered to contain between 6 and 7 per cent moisture, was used. The mold was "bottom poured," the bars being arranged so as to minimize the heating effects of the large bars on the small ones (Fig. 1) during cooling after casting.

6. The surfaces of the mold cavities were blackened with a mixture of water, graphite, clay and molasses to produce castings

¹ Moldenke, R., "Principles of Iron Founding," 2nd edition. McGraw-Hill Book Co., 1930, p. 243.

² Goerens, P., "Ueber die Konstitution des Roheisens." *Stahl und Eisen*, v. 26, April 1, 1906, p. 397.

³ TESTING AND GRADING FOUNDRY SANDS, 1931 edition. American Foundrymen's Association, p. 136.

free from "burnt-on" sand. The cheek and drag were then dried for 7 hours at 260 degs. Cent. (500 degs. Fahr.), and the castings were made on the following day. The cope, containing the pouring basin and made of green Lumberton sand, was placed in position just before the iron was poured.

7. Molds for casting bars for determining the linear contraction, and spirals for determining the running qualities of the

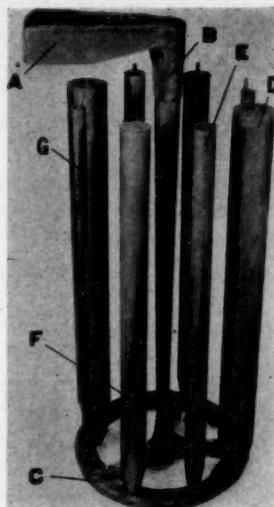


FIG. 1—CASTING SHOWING THE METHOD OF POURING BARS FOR TRANSVERSE TEST. A: POURING BASIN. B: DOWN GATE. C: FEEDING RING. D: 2.2-INCH DIAMETER BAR. E: 1.2 INCH DIAMETER BAR. F: 1.5-INCH DIAMETER BAR. G: 0.75-INCH DIAMETER BAR.

metal, were made of No. 00 Albany green sand, tempered to contain 6 to 7 per cent moisture.

8. *Melting and Pouring.* The irons were melted in a high-frequency induction furnace of the tilting type in a crucible of commercial magnesia. A charge of approximately 225 lbs. of metal was heated to the desired temperature. The maximum heating temperatures used were 1400, 1500, 1600 and 1700 degs. Cent. (2550, 2730, 2910, and 3090 degs. Fahr., respectively), for each type of iron. The metal was held at this temperature for approximately one minute. The power input was then regulated so as to obtain the desired rate of cooling of the melt. The characteris-

ties of the furnace are such that the liquid metal continued to circulate as long as the power was on.

9. Samples for the determination of shrinkage were taken and spirals for determining the running properties were poured at several arbitrarily chosen temperatures as the metal cooled. The transverse bar mold was poured at a temperature approximately 150 degs. Cent. (300 deg. Fahr.) above the liquidus temperature of stock pig iron. A $\frac{3}{4}$ -inch and a $1\frac{1}{2}$ -inch linear shrinkage bar, 12 inches long, designed to give ample draft for removing the pattern from the sand without rapping, were cast prior to casting the transverse test bars. All molds were poured directly from the furnace.

10. *Temperature Measurements.* Temperatures up to 1600 degs. Cent. (2910 degs. Fahr.) were measured by a platinum to platinum-10 per cent rhodium couple protected with a glazed porcelain tube inserted into a 1-inch diameter graphite tube. The portion of the graphite tube immersed in the iron was coated with a layer of aluminum oxide cement approximately $1/16$ -inch in thickness.

11. An optical pyrometer was used for measuring temperatures above 1600 degs Cent. (2910 degs. Fahr.). In such instances, the temperatures indicated by the thermocouple and optical pyrometer were read simultaneously from 1400 to 1600 degs. Cent. (2550 to 2190 degs. Fahr.), in order to check the optical pyrometer readings. The readings with the optical pyrometer were then continued up to 1700 degs. Cent. (3090 degs. Fahr.).

3—Transverse Test Bars.

12. The set of eight 23-inch length transverse test bars which were cast in one mold consisted of two bars of each of the diameters 0.75, 1.2, 1.5 and 2.2 inches. The bars cast to 1.5-inch diameter were machined to a diameter of 1.2 inches before testing to determine the "skin effect;" the others were tested as cast. The cross-sectional areas of the bars as tested were, therefore, in the approximate ratio of 1:3:9.

RESULTS AND DISCUSSION

1—Composition.

13. The results of the chemical analysis of each of the 12 heats are given in Table 1.

Table 1
CHEMICAL COMPOSITION AND PHYSICAL PROPERTIES OF TRANSVERSE TEST BARS CAST OR MACHINED TO 1.2 INCH DIAMETER.

Iron	Cast or Machined to Size	Max. Heating Temp., Degs. Cent.	Pouring Temp., Degs. Cent.	Degs. Cent.	Degs. Cent.	Chemical Composition, ¹ per cent						Temp. of Liquidus, Degs. Cent.	Temp. of Fahr.	
						TC	GC	CC	Si	S	Mn	P		
	Cast.....	1400	2550	1300	2370	3.57	2.81	0.76	2.65	0.021	0.20	0.86	1170	2140
	Machined.....	1400	2550	1300	2370	3.60	2.77	0.83	2.65	0.021	0.20	0.86	1170	2140
	Cast.....	1500	2730	1300	2370	3.67	2.99	0.68	2.44	0.014	0.23	0.85	1170	2140
A	Machined.....	1500	2730	1300	2370	3.67	2.99	0.68	2.44	0.014	0.23	0.85	1170	2140
	Cast.....	1600	2910	1300	2370	3.55	2.94	0.61	2.70	0.013	0.21	0.83	1155	2110
	Machined.....	1600	2910	1300	2370	3.52	2.96	0.56	2.70	0.013	0.21	0.83	1155	2110
	Cast.....	1700	3090	1300	2370	3.44	2.67	0.77	2.70	0.015	0.21	0.86	1175	2145
	Machined.....	1700	3090	1300	2370	3.46	2.66	0.80	2.70	0.015	0.21	0.86	1175	2145
	Cast.....	1400	2550	1380	2515	2.95	2.19	0.76	2.12	0.019	0.17	0.65	1225	2235
B	Cast.....	1500	2730	1380	2515	2.91	2.08	0.83	2.05	0.028	0.15	0.66	1240	2265
	Machined.....	1500	2730	1380	2515	2.90	2.09	0.81	2.05	0.028	0.15	0.66	1240	2265
	Cast.....	1600	2910	1380	2515	2.95	2.05	0.90	2.03	0.023	0.17	0.69	1240	2265
	Machined.....	1600	2910	1380	2515	2.95	2.06	0.89	2.03	0.023	0.17	0.69	1240	2265
	Cast.....	1700	3090	1380	2515	2.80	1.88	0.92	2.17	0.023	0.19	0.72	1245	2275
	Cast.....	1400	2550	1300	2370	3.84	3.12	0.72	1.17	0.054	0.54	0.10	1185	2165
C	Cast.....	1500	2730	1300	2370	3.76	2.66	1.10	1.16	0.055	0.57	0.10	1200	2230
	Cast.....	1600	2910	1300	2370	3.64	2.74	0.90	1.17	0.055	0.58	0.12	1220	2230
	Cast.....	1700	3090	1300	2370	3.60	2.76	0.84	1.16	0.054	0.62	0.10	1215	2220

Breaking Load cor- rected to 1.2 in. Diam., lbs.	Modulus of Rupture, lbs. per sq. in.	Moduli of Elasticity			Hardness Numbers			Density, gms. per cu. cm.	
		1200 lbs.	1500 lbs.	Ultimate	Brinell ²	Rockwell ³	Vickers ⁴		
1480	0.23	39,300	8.7	7.9	7.6	134	83	181	7.029
1320	0.27	35,100	6.4	5.9	131	78	170	7.080
2120	0.25	55,300	11.9	11.0	8.0	170	91	194	7.044
1720	0.28	45,400	9.9	7.6	163	80	170	7.056
1690	0.20	44,900	11.7	11.0	10.2	187	94	198	7.068
1600	0.23	42,600	10.3	8.5	174	90	194	7.061
1610	0.20	42,800	12.5	11.5	10.0	183	95	211	7.099
1510	0.27	40,100	7.7	6.7	167	90	182	7.078
2830	0.25	75,200	16.3	16.0	13.5	236	102	252	7.239
2670	0.23	70,800	16.5	16.0	13.9	241	99	248	7.282
2650	0.26	70,300	16.4	12.5	212	95	223	7.278
2710	0.27	71,500	15.0	14.5	12.1	229	97	236	7.285
2660	0.27	70,600	15.1	11.8	223	98	231	7.246
2780	0.22	73,900	18.2	17.8	15.3	248	104	257	7.278
2040	0.30	54,200	11.8	10.8	8.0	167	90	188	7.170
2280	0.22	60,600	19.3	18.0	12.7	174	95	195	7.189
2590	0.34	68,900	14.3	13.0	9.1	183	88	173	7.208
2620	0.28	68,500	19.3	18.5	11.2	192	95	207	7.214

¹Analyses made by R. H. Elder and Roy Deas, American Cast Iron Pipe Co., Birmingham, Ala.

²3000-kg. load, 10-mm. ball.

³B-scale, 100-kg. load, 1/16-inch ball.

⁴30-kg. load, 130 degree Diamond pyramid.

2—Shrinkage.

14. The shrinkage properties of the three types of iron, expressed as changes in specific volume with respect to temperature, are given in Fig. 2. The shrinkage of the iron in the liquid state was determined by the immersed-crucible method described by Saeger and Ash.⁴ The data on the shrinkage of the metal after

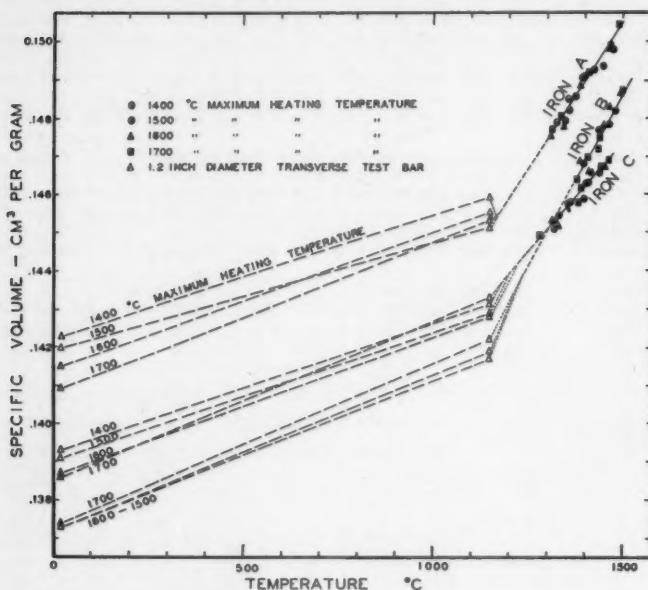


FIG. 2—SPECIFIC VOLUME-TEMPERATURE RELATIONS OF IRONS A, B AND C.

solidification were determined by the shop method described in a previous publication by Ash and Saeger.⁵

15. The value given for linear shrinkage is the average of the value obtained for the $\frac{3}{4}$ -inch and for the $1\frac{1}{2}$ -inch bar of the "no-rap" type. The specific volume of the metal at room temperature was determined on a sample cut from the 1.2-inch transverse bar as cast. The liquidus temperature for each iron was determined by thermal analysis (Table 1).

⁴ Saeger, C. M., Jr. and Ash, E. J., "A Method for Determining the Volume Changes Occurring in Metals During Casting," *Research Paper No. 399, Bureau of Standards Jnl. of Research*, v. 8, n. 1, 1932, p. 37.

⁵ Ash, E. J. and Saeger, C. M., Jr., *Shop Method for Determining Volume Changes in Cast Iron During Casting*, *TRANSACTIONS, A.F.A.* (1932), v. 40, p. 188.

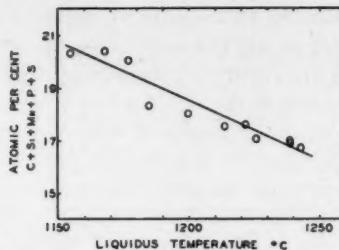


FIG. 3—RELATION OF LIQUIDUS TEMPERATURE TO ATOMIC PER CENTS OF CARBON, SILICON, PHOSPHORUS, MANGANESE AND SULPHUR.

16. Fig. 3 shows the relation between the composition of cast iron and its liquidus temperature. The sum of the percentages of carbon, silicon, manganese, sulphur and phosphorus have been expressed as atomic percentages of these elements.

17. The specific volume-temperature relation for the three irons in the liquid state were not affected by the heating temperature. The specific volume of irons *A* and *C* at room temperature decreased (density increased) with an increase in the heating temperature above the pouring temperature.

18. The density was determined by the conventional displacement-of-water method on the same specimens used for hardness determinations. The results are given in Fig. 4. Although

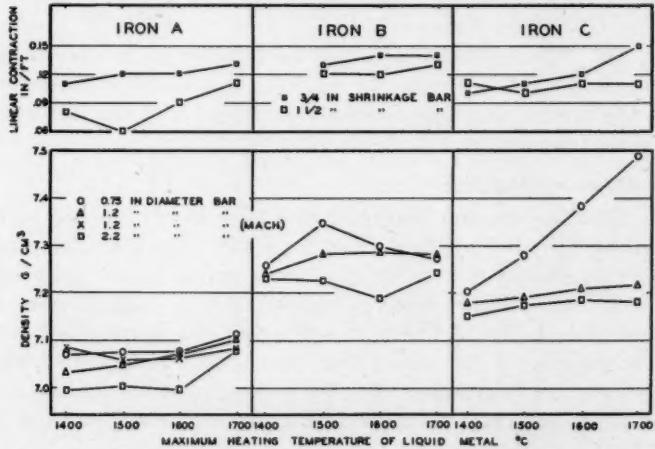


FIG. 4—VARIATIONS OF LINEAR SHRINKAGE AND DENSITY WITH MAXIMUM HEATING TEMPERATURE.

the differences obtained on the bars of different sizes were in excess of the differences attributable to the superheating prior to casting, in general, the density appears to increase with an increase in heating temperature of the liquid metal.

19. No data on the shrinkage of iron *B* were obtained for the heating temperature of 1400 degs. Cent. (2550 degs. Fahr.) on account of the narrow range between this temperature and the pour-

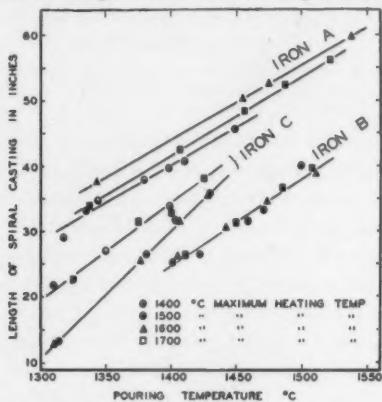


FIG. 5—RUNNING QUALITY OF IRONS *A*, *B* AND *C* AS Affected BY THE MAXIMUM HEATING TEMPERATURE.

ing temperature of the transverse bar mold, which was 1390 degs. Cent. (2535 degs. Fahr.).

20. The values for linear contraction for the individual bars are also plotted in Fig. 4. In general, these values as determined on the bars of both sizes, increased somewhat with the increase in heating temperature.

3—Running Qualities.

21. The running quality of each heat was determined by the method, described by Saeger and Krynnitsky,⁶ which consists essentially in casting a strip of small cross-section in the form of a spiral. Fig. 5 shows that the highest running qualities were obtained with the heats of iron *A*. Although iron *B* contained much more phosphorus and silicon than iron *C*, the running qualities of *C* were better than those of *B*. The running qualities of these irons are dependent upon the liquidus temperature, and apparently are

⁶ Saeger, C. M., Jr. and Krynnitsky, A. I., *A Practical Method for Studying the Running Qualities of a Metal Cast in Foundry Molds*. TRANSACTIONS A.F.A. (1931), v. 39, p. 513.

Table 2
COMPOSITION AND PHYSICAL PROPERTIES OF 0.75 INCH DIAMETER TRANSVERSE TEST BARS.

Maximum Heating Iron	Temperature Degs. Cent.	Pouring Temperature Degs. Fahr.	Degs. Cent.	Degs. Fahr.	Carbon Content, 1, 2 per cent.	0.75 in. Diam., lbs.	Breaking Load Corrected to 0.75 in. Diam., lbs.			Deflec- tion, inch	Modu- lus per eq. in.	Density, 7.066 7.072 7.074 7.109	Hardness Numbers— Vickers, ³ gms. per cu. cm.	
							TC	GC	CC		Rockwell, ⁴ in.	Brinell, ⁵ in.		
A	1400	2550	1300	2370	3.71 2.79 0.92	0.15	46,700	140	98	200
	1500	2730	1300	2370	3.69 2.90 0.73	510	0.15	60,300	201	98	226	7.072
B	1600	2910	1300	2370	3.63 2.90 0.63	660	0.18	212	99	245	7.074
	1700	3090	1300	2370	3.45 2.65 0.80	530	0.19	48,100	217	100	237	7.109
C	1400	2550	1380	2515	2.96 2.14 0.81	850	0.22	77,000	248	106	275	7.255
	1500	2730	1380	2515	2.88 2.08 0.80	870	0.23	79,000	245	103	288	7.347
	1600	2910	1380	2515	2.95 2.07 0.88	720	0.22	65,100	230	102	245	7.296
	1700	3090	1380	2515	2.82 1.99 0.83	800	0.20	73,000	262	106	286	7.371
	1400	2550	1300	2370	3.83 2.76 1.07	760	0.20	69,200	192	98	231	7.201
	1500	2730	1300	2370	3.77 2.61 1.26	810	0.20	74,000	217	102	239	7.277
	1600	2910	1300	2370	3.81 2.61 1.26	800	0.15	72,500	331	109	306	7.382
	1700	3090	1300	2370	3.62 1.34 2.28	800	0.15	302	...	380	7.468

¹Analyses made by R. H. Elder and Roy Dean, American Cast Iron Pipe Co., Birmingham, Ala.

²Complete composition given in Table 1.

³3000-lb. load, 10-mm. ball.

⁴B-scale, 100-kar. load, 1/16-inch ball.

⁵30-kar. load, 130 degrees Diamond pyramid.

Table 3
COMPOSITION AND PHYSICAL PROPERTIES OF 2.2 INCH DIAMETER TRANSVERSE TEST BARS.

Iron	Maximum Heating Temp., Degs. Cent.	Degs. Fahr.	Pouring Temperature, Degs. Cent., Degs. Fahr.	Carbon Content, per cent, TC ¹ GC ² CC ³	Modulus of Rupture, lbs. per sq. in.	Deflec- tion, inch	Hardness Numbers		Density, gms. per cu. cm.					
							Load Corrected to 2.2 in. Diam., lbs.	Brinell, ⁴ in.						
A	1400	2550	1300	2370	3.56	2.79	0.77	8,810	0.18	38,000	150	79	152	6.993
	1600	2730	1300	2370	3.66	3.00	0.66	10,110	0.18	43,600	145	83	166	7.000
	1600	2910	1300	2370	3.46	3.17	0.33	8,680	0.11	37,400	131	78	142	6.992
	1700	3090	1300	2370	3.46	2.68	0.73	8,450	0.15	36,400	151	90	180	7.073
B	1400	2550	1380	2615	2.92	2.20	0.72	14,200	0.14	61,200	213	95	203	7.236
	1600	2730	1380	2615	2.83	2.08	0.75	14,980	0.18	64,500	207	92	211	7.235
	1600	2910	1380	2615	2.91	2.09	0.82	13,640	0.15	68,700	201	92	219	7.187
	1700	3090	1380	2615	2.78	2.00	0.78	14,930	0.14	64,300	213	98	223	7.242
C	1400	2550	1300	2370	3.31	3.07	0.74	10,130	0.22	43,600	136	76	144	7.140
	1600	2730	1300	2370	3.73	2.77	0.96	11,720	0.20	50,500	152	82	149	7.171
	1600	2910	1300	2370	3.63	2.96	0.77	11,980	0.20	51,500	160	83	167	7.186
	1700	3090	1300	2370	3.59	2.84	0.75	11,700	0.16	50,700	162	83	168	7.180

¹Analyses made by R. H. Elder, American Cast Iron Pipe Co., Birmingham, Ala.

²Complete composition given in Table I.

³8000-kg. load, 10-mm ball.

⁴B-scale, 100-kg. load, 1/16-inch ball.

50-kg. load, 136 degrees Diamond pyramid.

not influenced by heating temperature above the pouring temperature. For a given pouring temperature, the heats of the iron having the lower liquidus temperature have superior running qualities.

4—Properties Determined on Transverse Test Bars.

22. *Breaking Load and Modulus of Rupture.* The breaking load was determined with an Amsler universal testing machine of 50,000 lbs. capacity, the load being applied to give a uniform rate of deflection for all bars, approximately 0.02 inch in 10 seconds. The span used in breaking the 1.2 and 2.2 inch bars was 18 inches, and a 15-inch span for the 0.75-inch bars.

23. The breaking loads, calculated to standard dimensions, for all sizes of bars are given in Tables 1, 2 and 3. Variations in diameter from 0.75 and 2.2 inches for cast bars of these nominal sizes were corrected by using correction factors derived on the same basis as those used for the 1.2-inch bars which are given in A.S.T.M. Specification A-124-29.⁷ Each value represents the average of the values obtained with the duplicate bars.

24. The modulus of rupture was calculated by means of the following formula:

$$F = 2.546 \frac{L \times D}{d^3}$$

where F = modulus of rupture,

L = breaking load in pounds,

D = distance between supports in inches; and

d = diameter of bar in inches.

25. Although any value of the modulus of rupture of cast iron is approximate, even for bars of the same length and section, the value may be regarded as a useful strength index, under standard testing conditions, of the combined influence of chemical composition, heating temperature, pouring temperature, structure, etc. Values of modulus of rupture of the three irons are shown in Fig. 6 and Tables 1, 2 and 3.

26. Tests on the 1.5-inch diameter bars, machined to 1.2-inch diameter, were completed only with iron *A*. The removal of a layer 0.15 inch thick from the surface of the bars resulted in values considerably lower than the corresponding values obtained on the unmachined 1.2-inch bars. The maximum strength for the four sizes of bars of iron *A* did not occur for the same maximum heating temperature. The same was found to be true for iron *B*.

⁷ A.S.T.M. Standards, 1930, pt. 1, p. 503.

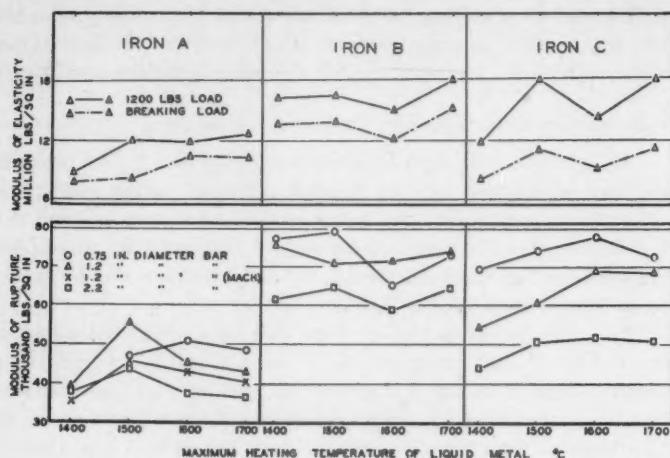


FIG. 6—MODULI OF RUPTURE AND MODULUS OF ELASTICITY AS AFFECTED BY MAXIMUM HEATING TEMPERATURE.

The heating temperature required to produce the highest strength for these two types of iron appears to be related to the cross-sectional area.

27. In the case of iron *C*, the maximum strength was obtained with the metal which had been heated to 1600 degs. Cent. (2910 degs. Fahr.) This result bears out the statement by Piwowarsky⁸ that, as the silicon content of a cast iron is lowered, the degree of superheat required to produce the maximum properties must also be raised.

28. On the basis of the results shown in Tables 1, 2 and 3, it would appear that irons *A* and *C* were much more susceptible to superheating than iron *B*. This iron practically did not respond.

29. *Deflection and Modulus of Elasticity.* Deflection measurements were measured by the travel of the bed of the Amsler machine during the loading period. The deflection-indicating device on the testing machine was found to check closely with a deflection meter whose least count was 0.001 inch. The deflections for all sizes of bars are given in Tables 1, 2 and 3.

30. The apparent modulus of elasticity for the 1.2-inch cast bars as affected by different degrees of superheating is shown in

⁸ Piwowarsky, E., *Progress in the Production of High-Test Cast Iron*. TRANSACTIONS A.F.A. (1926), v. 34, p. 937.

Fig. 6 and also in Table 1. The values of the modulus of elasticity given were calculated from the following formula:

$$E = \frac{L D^3}{48 I S}$$

where S = deflection,

L = load,

D = distance between supports,

I = moment of inertia; and

E = modulus of elasticity.

For round bars, the following calculation formula is used:

$$E = \frac{0.4244 \times L \times D^3}{d^4 S}$$

and when $D = 18$ inches and d = diameter of the bar, the following formula is used:

$$E = \frac{2475 \times L}{d^4 S}$$

31. This formula is based on assumptions which are not strictly true for cast iron, and the values obtained should not be compared with values for the modulus of elasticity of steels as ordinarily determined. It is believed, however, that the values are an index of the stiffness of the bars tested.

32. The moduli of elasticity calculated for the intermediate load of 1200 lbs. were much greater than the value calculated for the breaking load, and the values corresponding to the 1500-lb. load (Table 1) lay between the two. There is no obvious relationship between the moduli of rupture and of elasticity; that is, the higher values of the modulus of rupture do not correspond in many cases with the higher value of the modulus of elasticity.

33. *Hardness.* Hardness determinations were made by means of Brinell, Rockwell and Vickers hardness testing machines on discs $\frac{1}{2}$ -inch thick cut from the tested transverse bars adjacent to the fracture. The Vickers, Rockwell and Brinell numbers are given in Tables 1, 2 and 3, and are plotted in Fig. 7 for the various heating temperatures for the three irons.

34. Each of the Rockwell and Vickers numbers given are an average of seven readings, four near the circumference and three from the central portion of the specimen. One Brinell impression was made on the specimen from each of the 1.2 and 0.75 inch bars, whereas the numbers for the 2.2-inch bar are averages of five impressions.

35. The curves for hardness numbers of the different test specimens obtained for the different heating temperatures show

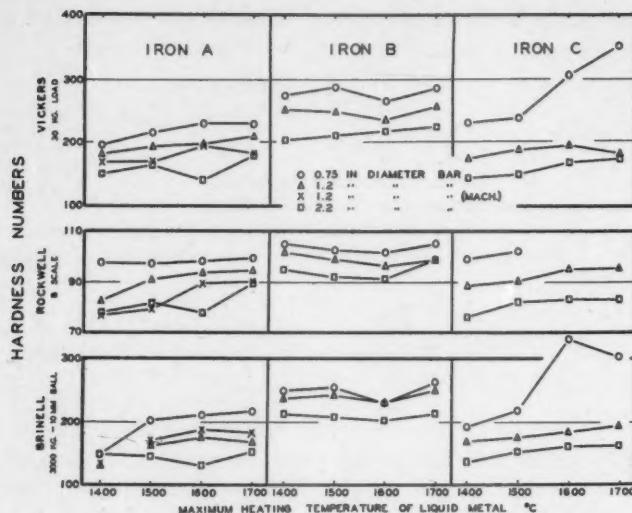


FIG. 7—RELATION OF HARDNESS TO MAXIMUM HEATING TEMPERATURE.

the same general trend as the corresponding curves representing the transverse breaking strength. Here again, the cross-sectional area of the cast bars is also a factor.

36. MICROSTRUCTURE. Cross-sections cut perpendicularly to the length of 1.2-inch bars from all the heats of the three irons (*A*, *B* and *C*) were examined microscopically. It is well known that the size and shape of the graphite particles have a pronounced effect on the physical properties of cast irons. The grinding and polishing of the specimens were done very carefully in order to preserve the graphite particles intact, so as to permit a comparison of graphite particles of different irons.

37. The method described by Vilella⁹ was employed in the preparation of the specimens, except that the final polishing was done by hand on heavy satin stretched tightly over a glass plate and covered with a thin layer of water-magnesia paste. This method was very satisfactory. Most of the graphite particles were retained intact in the polished specimens, as shown in Fig. 8-*A*.⁹

38. All micrographs shown in this paper represent the central portion of the 1.2-inch bars. The micrographs shown were

⁹ Vilella, J. R., "Improved Method of Polishing Metallographic Specimens of Cast Iron," *Metals and Alloys*, v. 3, 1932, p. 205.

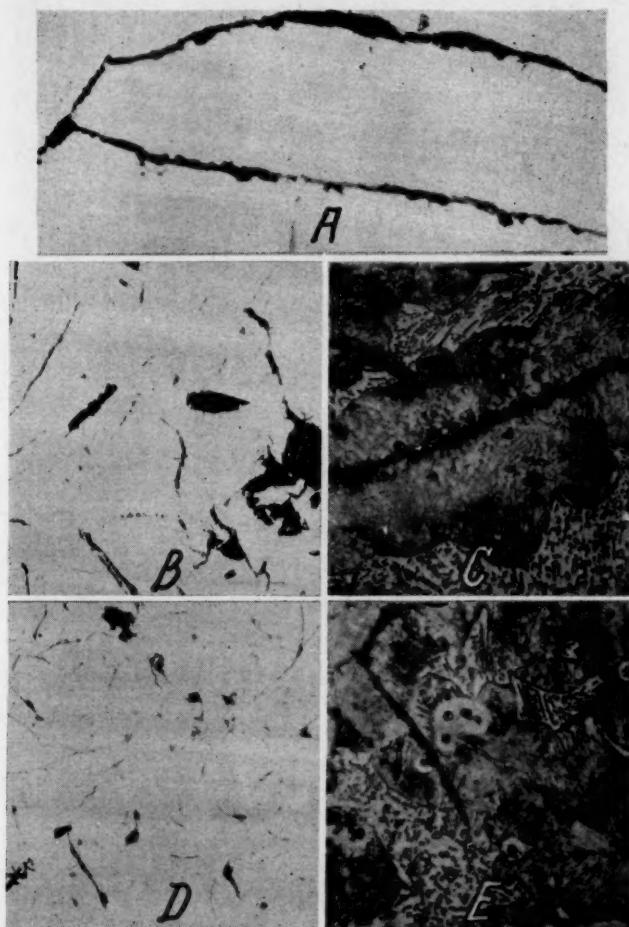


FIG. 8—MICROSTRUCTURE OF CENTRAL PORTION OF 1.2-INCH TRANSVERSE TEST BARS OF IRON A. ETCHED SPECIMENS SHOWN HERE AND IN FIGS. 9 AND 10, ETCHED IN 5% ALCOHOLIC SOLUTION OF PICRIC ACID (PICRAL). A: LOWEST STRENGTH IRON, MAXIMUM HEATING TEMPERATURE 1400 DEGS. CENT. (2550 DEGS. FAHR.); UNETCHED, X500. B: SAME AS A; X100. C: SAME AS A; ETCHED, X500. D: HIGHEST STRENGTH IRON, MAXIMUM HEATING TEMPERATURE 1500 DEGS. CENT. (2730 DEGS. FAHR.); UNETCHED, X100. E: SAME AS D; ETCHED, X500.

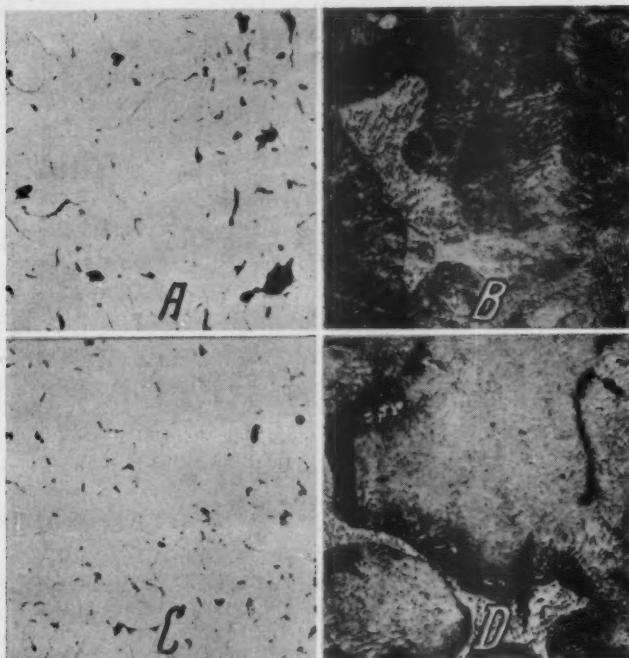


FIG. 9—CENTRAL PORTIONS OF 1.2-INCH TRANSVERSE TEST BARS OF IRON B.
 A: LOWEST STRENGTH IRON, MAXIMUM HEATING TEMPERATURE 1500 DEGS. CENT. (2730 DEGS. FAHR.); UNETCHED, X100. B: SAME SPECIMEN AS A; ETCHED, X500. C: HIGHEST STRENGTH IRON, MAXIMUM HEATING TEMPERATURE 1400 DEGS. CENT. (2550 DEGS. FAHR.); UNETCHED, X100. D: SAME SPECIMEN AS C; ETCHED, X500.

selected as representative of the structure of the 1.2-inch bars showing the highest and the lowest strength for each of the three irons.

39. Fig. 8 shows the microstructure of two samples of iron *A* which had been heated to temperatures of 1400 and 1500 degs. Cent. (2550 and 2730 degs. Fahr.) The outstanding structural features of iron *A* are the large amount of phosphide eutectic, the long, straight graphite flakes and the coarse pearlite. The structure of iron *A* having lowest strength, and which had been heated to a temperature of 1400 degs. Cent. (2250 degs. Fahr.), is shown in micrographs *B* and *C*, Fig. 8. The cavities and graphite flakes are very noticeably larger, and the pearlite is coarser than the corresponding structure of iron *A* having the highest strength,

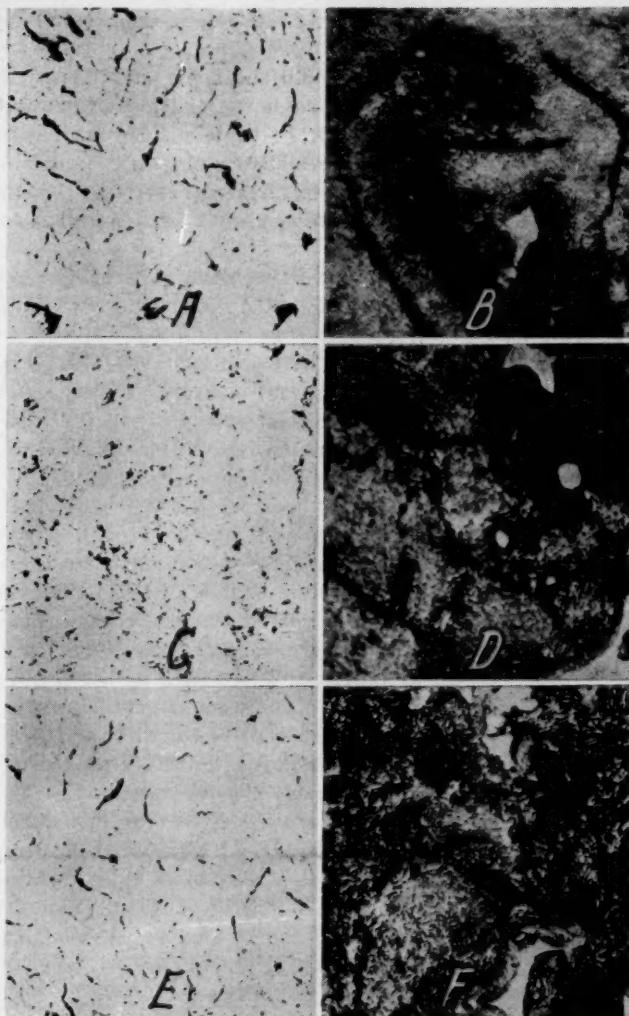


FIG. 10—CENTRAL PORTIONS OF 1.2-INCH TRANSVERSE TEST BARS OF IRON *C*. *A*: LOWEST STRENGTH IRON, MAXIMUM HEATING TEMPERATURE 1400 DEGS. CENT. (2550 DEGS. FAHR.); UNETCHED, X100. *B*: SAME AS *A*; ETCHED, X500. *C*: IRON POSSESSING VERY HIGH MODULUS OF ELASTICITY, MAXIMUM HEATING TEMPERATURE 1500 DEGS. CENT. (2730 DEGS. FAHR.); UNETCHED, X100. *D*: SAME AS *C*; ETCHED, X500. *E*: HIGHEST STRENGTH IRON, MAXIMUM HEATING TEMPERATURE 1600 DEGS. CENT. (2010 DEGS. FAHR.); UNETCHED, X100. *F*: SAME AS *E*; ETCHED, X500.

heated to a maximum temperature of 1500 degs. Cent. (2730 degs. Fahr.)

40. The difference in strength, hardness and density between the bars of iron *A*, having the lowest and highest strengths, was very pronounced and cannot be explained on the basis of chemical composition, because the highest strength iron contained a larger amount of graphite and a smaller quantity of combined carbon. The difference in microstructure affords a satisfactory explanation of this difference.

41. Fig. 9 shows the structure representative of two heats of iron *B*. In iron *B* a great amount of phosphide eutectic was also present. The lowest strength iron of this class (micrographs *A* and *B*, Fig. 9) which had been heated to 1500 degs. Cent. (2730 degs. Fahr.) had a pearlite structure associated with the coarser graphite. Micrographs of the highest strength bars of iron *B* which had been heated to a maximum temperature of 1400 degs. Cent. (2250 degs. Fahr.) are shown in *C* and *D* of Fig. 9.

42. Fig. 10 represents three heats of iron *C*. The sizes of graphite flakes shown in micrographs *A* and *E*, Fig. 10 of specimens from two heats which had been treated to maximum heating temperatures of 1400 and 1600 degs. Cent. (2250 and 2910 degs. Fahr., respectively), were quite similar to those in irons of class *B*.

43. The microstructure of the bars of iron *C* having the lowest strength, shown by micrographs *A* and *B* of Fig. 10, indicated that the graphite flakes were somewhat larger than those in the corresponding bars of highest strength, shown in *E* and *F*, Fig. 10. The structure of the highest strength iron consisted of a pearlitic-sorbitic matrix with islands of free cementite.

44. The graphite particles of iron *C*, heated to maximum temperatures of 1500 and 1700 degs. Cent. (2730 and 3090 degs. Fahr.) were much finer than those of any other irons examined. The structure of the iron heated to 1500 degs. Cent. (2730 degs. Fahr.) is shown in micrographs *C* and *D*, Fig. 10. It is of interest to note that the modulus of elasticity (Fig. 6 and Table 1) of these two irons was the highest of all observed.

IV—SUMMARY

45. The effects of maximum heating temperature of the liquid metal on the shrinkage, running qualities, transverse strength, hardness, density and microstructure of three types of iron are discussed.

46. The specific volume-temperature relations for the three irons in the liquid state were not affected by the maximum heating temperature. In general, the density and linear contraction increased with the degree of superheat imparted to the liquid metal.

47. The running qualities of these irons were dependent upon the liquidus temperature and apparently were not influenced by maximum heating temperature. For a chosen pouring temperature, the heats of iron having the lower liquidus temperature have superior running qualities.

48. The maximum strength for the bars of different diameters, for two types of iron, did not occur for the same maximum heating temperature and appeared to be influenced by the cross-sectional area. The strength of all bars for the third type of iron varied proportionally with the maximum heating temperature.

49. Tests on the 1.5-inch diameter bars machined to 1.2-inch diameter indicated that the removal of the skin layer resulted in values considerably lower than the corresponding values obtained on the 1.2-inch diameter bars as cast. The deflection value for the machined bar was, in general, greater than the deflection of the 1.2-inch bar as cast.

50. There is no obvious relationship between the moduli of rupture and of elasticity; that is, the higher values of the modulus of rupture do not correspond in many cases with the higher values of the modulus of elasticity.

51. Brinell, Rockwell and Vickers hardness numbers for the different maximum heating temperatures show the same general trend as the modulus of rupture.

52. The cross-sectional areas of the cast bars appear to be the controlling factor for strength and hardness, the effect of superheating prior to pouring being a relatively minor one.

53. The microstructure of the 1.2-inch bars indicates that those irons of low strength are associated with relatively large straight graphite flakes and coarse pearlite, whereas the irons of highest strength are associated with relatively small graphite flakes and a matrix of fine pearlite or sorbite.

V—Acknowledgment

54. The authors are indebted to A. I. Krynnitsky, who prepared the micrographs and otherwise rendered valuable assistance and advice in this investigation, and also acknowledge their in-

debtors to the American Cast Iron Pipe Co. for making the chemical analyses.

ORAL DISCUSSION

J. T. MACKENZIE:¹ I would like to point out, in connection with electric furnace operation, that the carbon of iron *A*, melted in the induction furnace with a magnesia crucible with only the oxygen of the air present, dropped from 3.67 to 3.44 per cent, on iron *B* it dropped from 2.95 to 2.80 per cent, and on iron *C* it dropped from 3.84 to 3.60 per cent. From this we might accept the fact that this drop in carbon is more a function of the maximum heating temperature of cast iron than it is of the type of the electric furnace.

MEMBER: Is there a critical temperature at which the cast irons may be poured that will bring out the better physical properties of the iron?

J. T. MACKENZIE: The authors did not investigate the pouring temperatures. Temperatures given are maximum heating temperatures. They endeavored to pour test bar molds at 270 degrees Fahr. (150 degrees Cent.) above the liquidus temperature no matter what maximum heating temperature was being investigated.

CHAIRMAN R. S. MACPHERRAN:² I believe the most important point brought out is that some irons are benefited definitely by superheating and some are injured. Superheating is not a good thing for all irons. Low carbon irons are injured by superheating.

¹ American Cast Iron Pipe Co., Birmingham, Ala.

² Allis-Chalmers Mfg. Co., Milwaukee.

Silicon and Manganese as Deoxidizers in Cast Iron

BY A. H. DIERKER,* COLUMBUS, O.

Abstract

To determine the effects of varying the amounts of manganese on the physical properties of a gray cast iron, manganese was added to a low manganese iron, producing specimens with 0.46, 0.62, 0.94 and 1.20 per cent manganese. Brinell hardness dropped on the first addition and then gradually increased on further additions. The original pearlitic structure was changed first to a ferritic structure. As more manganese was added, the amount of pearlite was increased in proportion to the manganese added. The author presents his theory to account for these changes.

1. In certain melts of gray cast iron made in the laboratories of the Engineering Experiment Station, Ohio State University, Columbus, O., a marked change in physical properties was secured by making increasing additions of small amounts of manganese.

2. Fig. 1, curves 1 and 2, shows the effect on Brinell hardness of increasing additions of manganese to iron low in that element. The specimens were all poured from the same melt of the following composition: TC., 3.24 per cent; Si., 1.93 per cent; P., 0.02 per cent; S., 0.02 per cent; and Mn., 0.18 per cent. (Fig. 6 shows the type of specimens used.) This melt was made in an electric arc furnace. No manganese was added before pouring the first set of specimens. Increasing manganese in the remaining specimens was secured by ladle additions of 80 per cent ferromanganese.

3. It will be noted that the drop in hardness on the first man-

* Engineering Experiment Station, Ohio State University.

NOTE: This paper was presented at one of the sessions on Cast Iron at the 1933 Convention of the American Foundrymen's Association.

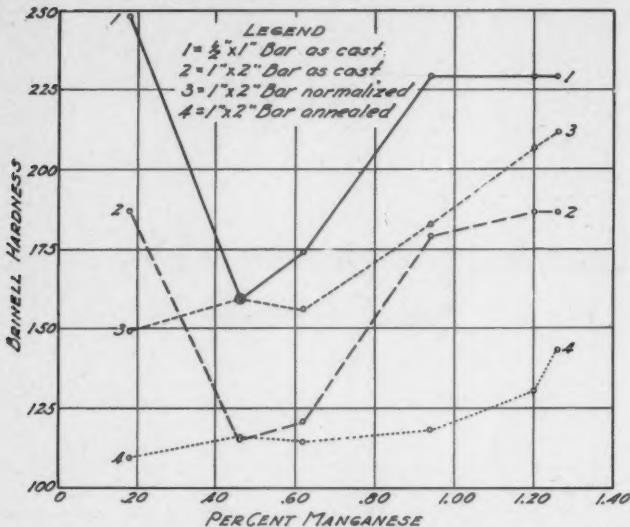


FIG. 1—EFFECTS ON BRINELL HARDNESS OF INCREASING ADDITIONS OF MANGANESE TO IRON OF LOW MANGANESE CONTENT.

ganese addition is abrupt and that increasing manganese gradually increased the hardness. With the change in hardness, the specimens showed a corresponding change in structure which can be seen from an inspection of the micrographs shown in Fig. 2. It will be noted that the original structure is pearlitic. The first manganese addition changes the structure to an almost completely ferritic structure. With further manganese additions, there is a gradual return to the pearlitic structure.

4. Other investigators have found previously that small increases of manganese in low sulphur, low manganese irons had an apparent graphitizing effect. Norbury¹ explains the graphitizing action of small amounts of manganese as an indirect one due to manganese combining with and nullifying the carbide-stabilizing action of traces of iron sulfide. Apart from this indirect graphitizing action, Norbury found the normal function of manganese is to stabilize iron carbide and to increase the tendency to chill.

5. This would appear a reasonable explanation of the phenomena described except that in this case the residual manganese present seems ample to take care of the small amount of sulphur.

¹ A. L. Norbury, *The Influence of Manganese in Cast Iron*, Proceedings, Institute of British Foundrymen, Vol. 22, pp. 151-176, 1928-1929.

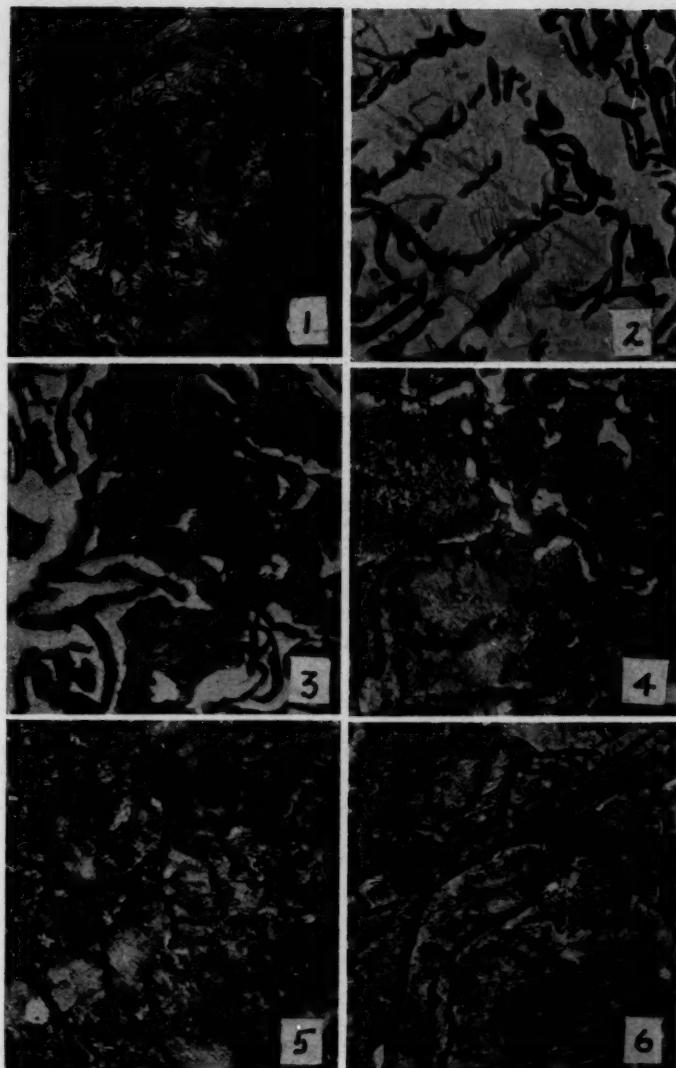


FIG. 2—MICROGRAPHS SHOWING CHANGE IN STRUCTURE OF AN IRON WITH INCREASING MN ADDITIONS. MICROGRAPHS TAKEN AT CENTER OF 1X2-IN. BAR AS CAST, 200X. (1) 0.18 PER CENT MN; (2) 0.46 PER CENT MN; (3) 0.62 PER CENT MN; (4) 0.94 PER CENT MN; (5) 1.20 PER CENT MN; (6) 1.26 PER CENT MN.

The effect of normalizing and annealing on the specimens also leads away from the sulphur theory. Curves 3 and 4, Fig. 1, show that on either normalizing or annealing, the specimens assume their logical order; *i.e.*, increasing hardness with increasing manganese content.

6. By assuming oxygen to be an important element in the above specimens, it is possible to evolve a theory that, although not conclusive, seems to fit the facts in the case.

7. The principal deoxidizing elements in cast iron are carbon, manganese, and silicon. The primary reactions are:

1. $\text{Si} + 2 \text{FeO} = \text{SiO}_2 + 2 \text{Fe} + 65 \text{ Cals.}$
2. $\text{Mn} + \text{FeO} = \text{MnO} + \text{Fe} + 26 \text{ Cals.}$
3. $\text{C} + \text{FeO} = \text{CO} + \text{Fe} - 36 \text{ Cals.}$

8. These reactions are reversible and do not go to completion. They tend to reach equilibrium and, theoretically, the melt can never be deoxidized completely. The equilibrium constant will vary with the temperature. At elevated temperatures, reactions 1 and 2, being exothermic, will tend to move toward the left while reaction 3, being endothermic, will move toward the right. It is an accepted fact that manganese and silicon are most effective as

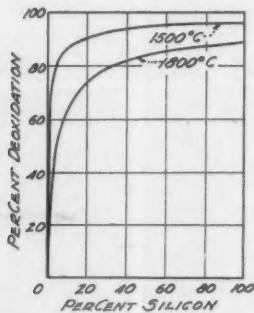


FIG. 3—EFFECT OF SILICON AS A DEOXIDIZER (McCANCE).

deoxidizers at relatively low temperatures, while carbon is best at high temperatures.

9. Just how much oxygen may be present in molten cast iron is not known. However, it is probable that any cast iron melted by ordinary commercial methods will contain oxygen in sufficient

amount to effect the physical properties of the material when solidified. Assuming that some FeO is present in the superheated molten metal, it is possible to visualize the reactions that take place as the temperature drops.

10. Fig. 3 is taken from the work of McCance.² Attention is called to it merely to show the tendency of silicon as a deoxidizer. It can be seen readily that, as the temperature drops, silicon

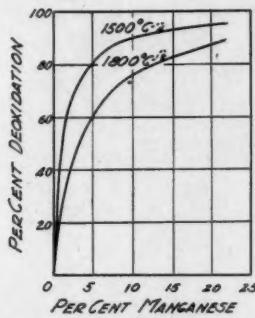


FIG. 4—EFFECT OF MANGANESE AS A DEOXIDIZER.

becomes more effective as a deoxidizer. This would be suspected, since the oxidation reaction is exothermic. As the temperature drops, more and more silicon is oxidized. McCance believes that the SiO_2 formed is precipitated as fine, dust-like particles of submicroscopic size which, due to their high melting point and viscosity, do not coagulate readily. This precipitate easily could be a ferrous silicate high in silica instead of straight SiO_2 . In either case, the effect would be the same.

11. Fig. 4 shows that manganese, like silicon, becomes more effective as a deoxidizer as the temperature drops and for the same reason. With both silicon and manganese present in the melt, the amount of deoxidation accomplished by each is dependent on the relative concentration of the two elements. MnO is quite basic, is soluble in FeO and probably in the melt. It combines readily with SiO_2 and FeO to form iron manganese silicates. Herty³ has found that silicates high in MnO apparently coagulate into larger sized particles than those low in this oxide. Although there may

² A. McCance, *Balanced Reactions in Steel Manufacture*, Transactions, Faraday Society, Vol. 21, p. 176, 1925.

³ C. H. Herty, Jr., *Fundamental and Applied Research on the Physical Chemistry of Steel Making*, U. S. Bureau of Mines, R. I. 3054, 1930.

be some doubt as to whether or not these are formed in time to levitate out of the melt, still, for reasons to be discussed shortly, it is believed that these larger particles have a markedly less effect on the properties of the iron in which they occur, than the finely disseminated siliceous particles to which previous reference has been made.

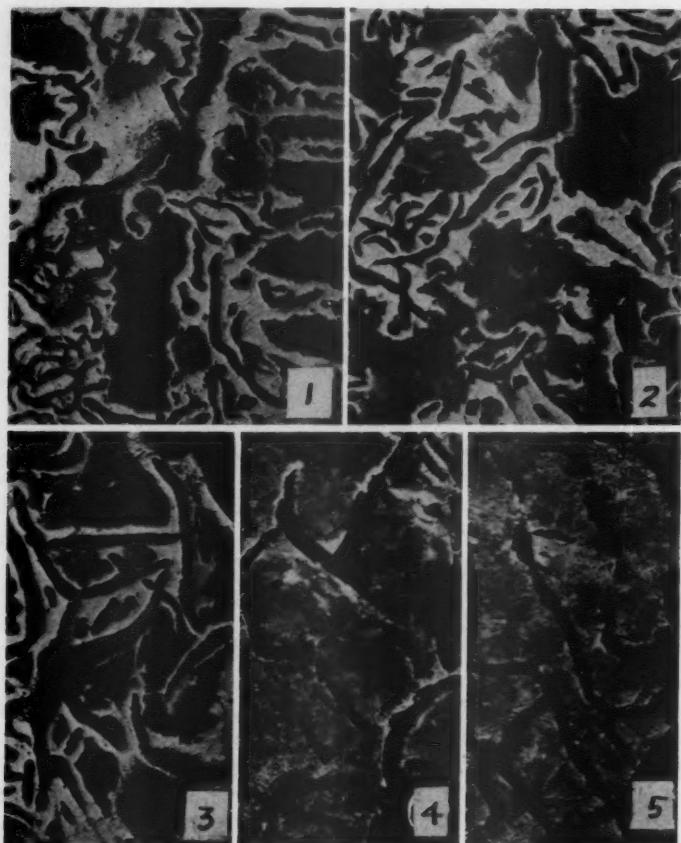


FIG. 5—MICROGRAPHS SHOWING CHANGE IN STRUCTURE OF NORMALIZED IRONS WITH INCREASING MANGANESE ADDITIONS. MICROGRAPHS TAKEN AT CENTER OF 1x2-IN. BAR, 200X. (1) 0.18 PER CENT MANGANESE; (2) 0.46 PER CENT MANGANESE; (3) 0.62 PER CENT MANGANESE; (4) 0.94 PER CENT MANGANESE; (5) 1.20 PER CENT MANGANESE.

12. In the chemical analysis of cast iron for silicon, no distinction is made between the silicon present in solid solution and that present as an oxide. Hence, all the silicon shown by analysis may not necessarily be in solid solution and therefore may not all be available as a graphitizer. It would appear unlikely that sufficient silicon was oxidized to account for the wide difference in graphitization noted in the experiments described if it is merely a matter of taking silicon out of solution. It is not unreasonable to expect that, if the oxidized silicon is precipitated as a cloud of

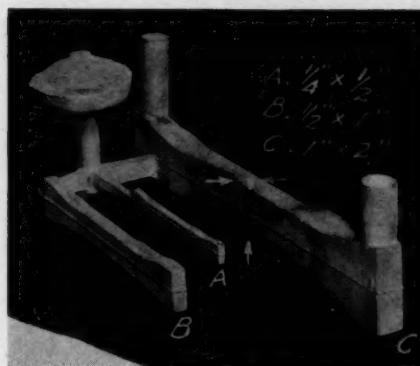


FIG. 6—SIZES AND SHAPES OF THREE TEST BARS.

submicroscopic particles, these particles could obstruct the diffusion of carbides to the grain boundaries and cause these carbides to precipitate out within the grain itself forming pearlite. Coe⁴ found that manganese also has a marked effect on the diffusive power of iron carbide and, with sufficient manganese present, instead of being driven to the edges of the solid solution, the carbide separates in the solid solution itself.

13. On this basis, an attempt will be made to explain the structures shown in Fig. 2. In the first specimen, due to oxidation of some of the silicon and to the nature and distribution of the precipitated oxides, the carbides are prevented from diffusing to the grain boundaries and the pearlitic structure shown results. In the second specimen, not only has less silicon been oxidized due to the addition of manganese, but the oxides formed are of a dif-

⁴ H. I. Coe, *Manganese in Cast Iron and the Volume Changes During Cooling*, *Journal, Iron and Steel Institute*, Vol. 82, Pt. II, p. 105, 1910.

ferent type, a type which does not hinder carbide diffusion. The carbides readily diffuse to the grain boundaries and graphitize. Thus the ferritic structure shown is secured. As more manganese is added, that element becomes an active agent in obstructing carbide diffusion and the pearlitic structure again is present.

14. All these structures have been secured by cooling from the melt. In reheating, the oxidation reactions are reversed and if, in the case of the first specimen, the silica is as finely divided as

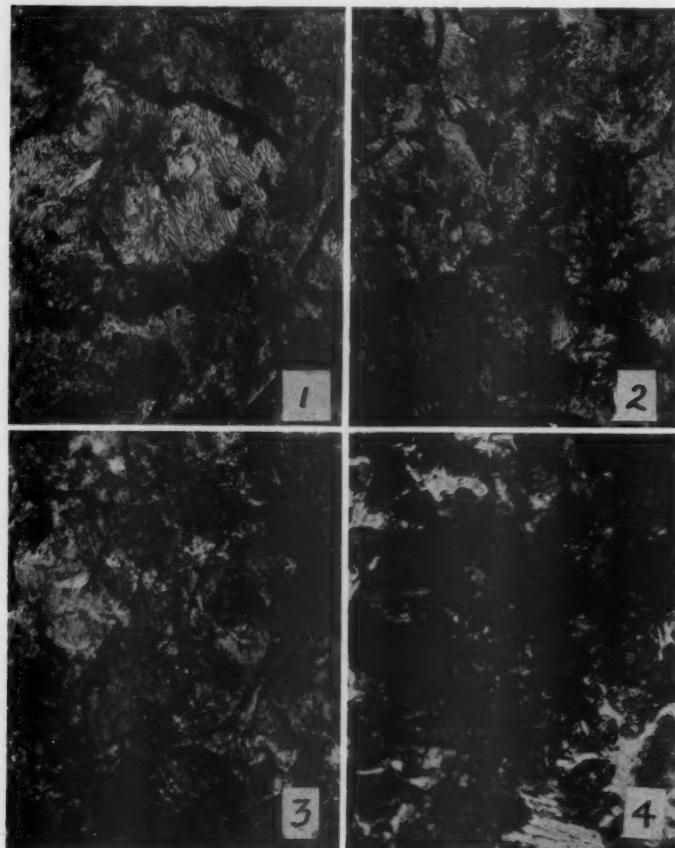


FIG. 7—STRUCTURE OF IRONS OF THE SAME COMPOSITION CAST INTO BARS OF VARIOUS SIZES. (1) $1\frac{1}{2} \times 3$ -IN.; (2) 1x2-IN.; (3) $\frac{1}{2} \times 1$ -IN.; (4) $\frac{1}{4} \times \frac{1}{2}$ -IN. STRUCTURE AT CENTER OF EACH BAR, 200X.

it is assumed to be, it is possible for it to be reduced by the carbon with an increase in FeO content; *i.e.*, the O₂ goes back in solution. Grossman⁵ and Herty⁶ have found that in carburized steels low in manganese, dissolved oxygen caused abnormality; *i.e.*, diffusion and concentration of the carbides leaving ferrite areas, between the pearlite and the grain boundary carbides.

15. In gray cast iron, the carbides are unstable and graphitize. Therefore, instead of grain boundary cementite, graphite flakes are present. If the first specimen is low in manganese and

Table 1
TESTS OF GRAY CAST IRON CONTAINING 0.40 TO 0.50 PER CENT MANGANESE

Size of Specimen	Tensile Strength Lbs. 1 sq. in.	Brinell No.	Composition of All Specimens
1½ in. x 3 in.	38,650	203	TC 3.12
1 in. x 2 in.	40,550	213	Si 1.26
½ in. x 1 in.	42,800	255	P 0.024
¼ in. x ½ in.	77,500	363	S 0.040
			Mn. 0.48

contains dissolved oxygen, we would expect to find, on normalizing, ferrite areas surrounding the pearlite. We also would expect to find the free ferrite lessened as the manganese increased. Fig. 5 shows specimens to be as expected with the free ferrite areas decreasing with increasing manganese content.

16. Although having no direct bearing on the arguments presented, it may be of some interest to conclude the discussion by showing the results of tests of a melt of gray iron in which it was attempted to keep the residual manganese at a point where maximum graphitization in the material would result; *i.e.*, 0.40 to 0.50 per cent manganese. The melt was made in an electric furnace and poured into bars of the type shown in Fig. 6. The largest bar was poured separately and is not shown. These bars, although perhaps not acceptable as a standard, in general have proved satisfactory for the type of work in which the effect of size of specimen on structure is of more importance than an absolute measure of tensile strength. They are easily cast, easily finished, and need no special equipment for testing. Table 1 shows the results of tests of these bars poured from the melt mentioned. All speci-

⁵ M. A. Grossman, *Oxygen in Steel. Transactions, A.S.S.T.*, Vol. 18, p. 601.

⁶ C. H. Herty, Jr., *Non-Metallic Inclusions in Steel. Transactions, A.S.S.T.*, Vol. 19, p. 36.

mens except the smallest, which was quite hard, were machinable. Fig. 7 shows the structures, which are pearlitic throughout except the last, which showed large carbide areas.

ORAL DISCUSSION

MEMBER: Can Mr. Dierker tell us the percentage of steel in his base charge?

A. H. DIERKER: I do not recall the exact makeup of the base charge. We used a small direct arc furnace on these melts and the charges were made up entirely of scrap from previous melts. Some of these previous melts were a synthetic iron of steel plus graphite. Where a melt extremely low in manganese, phosphorus and sulphur is wanted, we use special Armco melting scrap plus graphite. We have also used wash metal. There was no steel at all in the heats of these test bars, just scrap from previous melts.

MEMBER: Did you duplicate those results, pouring the test bars one to a flask rather than as a series in one flask? The heat in the pouring box will affect the physical properties and the microstructure, provided the bars are all poured on the same sprue.

A. H. DIERKER: We have not attempted the separate casting method. We pour sets of castings in the same way and get a comparison between one melt and another. I do not think you can use those results as an absolute measure and we do not give them out as such.

R. S. MACPHERRAN:¹ Did the author determine the amounts of oxygen in the specimens shown in micrographs 1 and 2, Fig. 2?

A. H. DIERKER: We made no attempt to determine the oxygen, for several reasons. It is a difficult determination to make and you are never sure that it has been done exactly right. In this particular instance, it would have proved nothing if we had found exactly the same amount of oxygen present in each case. If you will recall my arguments, it is not the amount of oxygen present but the form the oxygen takes that is important. Had we determined the total amount of oxygen, it might have shown the same in either case, and our results would have been of no value. Since we have been unable to find any method of determining the exact form which the oxygen takes, we have to theorize until we do.

R. S. MACPHERRAN: Various determinations made years ago by Johnson and Stewart pretty well proved that the presence of oxygen in small amounts will both harden and strengthen cast iron. The removal of that oxygen might explain the change in samples 1 and 2, Fig. 2. We have found in chill blocks that, if the manganese is low, the addition of a small amount will take out part of the chill. When a certain limit is reached the manganese increases the depth of chill. I wonder whether this effect was not due partly to a change in the amount of oxygen present.

¹ Allis-Chalmers Mfg. Co., South Milwaukee, Wis.

A. H. DIERKER: Johnson's work was very good except that I think he overlooked certain things when adding oxygen to cast iron in any quantity, especially in the quantities which he added where he actually blew his metal by a sort of a modified Bessemer process. In the Bessemer process, they burn out the silicon and carbon and that is exactly what Johnson did. When that quantity of oxygen is added, you actually oxidize out a very appreciable amount of the silicon. He lowered his silicon and carbon content and, of course, he made harder and stronger iron. He put oxygen in deliberately but his analysis showed he oxidized out an appreciable amount of his silicon and carbon.

H. B. SWAN:² In reference to Johnson's work, I happened to have done some of his work at our plant. As I recall it, there was no change in either the silicon or manganese, and this particular addition of strength was maintained through repeated meltings of that iron. I do not recall that the additional strength as checked up with the micrographs was due to oxidation of either the manganese or silicon.

A. H. DIERKER: Mr. Johnson,³ in reporting his work states that "A heat of metal of 1.0 per cent silicon was blown in a side blow convertor . . . until the carbon flame began to break through, showing the silicon was gone. . . . With this metal was mixed an equal volume of 2.0 per cent silicon iron direct from the cupola and from that mixture test bars were cast."

Mr. Swan undoubtedly refers to crucible remelts of this material made at the Cadillac Motor Car Company, in which case, there would be no reduction in silicon.

H. BORNSTEIN:⁴ Mr. Dierker, did you make any further experiments with more commercial types of iron? We know that 0.18 per cent manganese is not commercial, neither is .02 per cent sulphur. I was wondering whether you had made some tests, starting in with a commercial analysis and then making manganese additions to note the effect. We would expect that an iron of the composition shown, with about 0.50 per cent manganese, would not have ordinarily a ferritic structure such as you obtain by the manganese additions. Suppose you start with a commercial analysis, what would you obtain later?

A. H. DIERKER: We have not gone that far as yet. We happen to be in the fortunate position that we do not have to get commercial results, but we like to do it when we can carry experiments that far. Of course, you understand that when you get into commercial cast iron in the high sulphur and high phosphorus class, you are introducing certain other variables. When we start, we like to keep the variables as few as possible, then gradually increase as we go along. We hope to carry on experiments into the commercial ranges and, when we do, we will be glad to report our results.

H. BORNSTEIN: I cannot see that you are introducing any more variables. You are introducing different quantities of those variables.

² Cadillac Motor Car Co., Detroit, Mich.

³ *Met. & Chem. Eng.*, Vol. XV, No. 11, p. 644.

⁴ Deere & Co., Moline, Ill.

You still have the sulphur, whether you have 0.02 or 0.06 per cent. You still have the manganese whether it is 0.18 or 0.50 per cent.

MEMBER: Did you notice any kick-back in the risers upon the addition of increasing manganese, a sort of blow-back in the risers in any of your tests?

A. H. DIERKER: We noticed nothing abnormal in the pouring of the bars. They all had normal shrinkage and all were solid at the gates. Had there been a difference, we probably would have noticed it.

The Field for Materials Handling in Semi-Production Foundries

By RALPH G. WIELAND,* CLEVELAND

1. This paper has been prepared to stimulate the design of materials-handling equipment for semi-production foundries. As is well known, the production foundries such as are engaged in the manufacture of automotive parts, pipe, sanitary, boilers, radiators, etc., have cut their costs tremendously through the use of materials-handling methods.

2. Comparatively few semi-production foundries, however, have installed machinery to handle their materials more efficiently. This is due to the wide range of products cast in a single shop, since such shops are mainly jobbing foundries, and also to the fact that materials-handling equipment manufacturers have, in the past, devoted most of their efforts to the design of equipment for large-production, continuous-pour foundries.

3. It will be the object of this paper, therefore, to outline some of the fundamentals for better methods of materials handling in semi-production foundries.

4. The semi-production foundry may be defined as a plant melting from 10 to 50 tons per day but which is not operated as a continuous-pour shop. Such a foundry may also be described as employing from 10 to 70 molders, working chiefly on power molding machines with orders from 100 molds to 2000 molds on individual patterns. Some molders may be running two or three different patterns on the same day in different size flasks; some work will be cored, and some will be plain. Casting weights may vary from one pound to 300 lbs., and sections may vary from $\frac{1}{8}$ inch to 2 inches in thickness.

5. While it is desirable at all times to produce the smoothest castings possible, in a shop of this type there will be a consider-

* Forest City Foundries Co., Plant "B."

Note: This paper was presented before the Materials Handling session at the 1933 Convention of American Foundrymen's Association.

able difference in the required finish of castings from different patterns.

6. From the above description it will be seen that, while this paper will apply chiefly to the gray iron foundry, part of the text nevertheless will be equally applicable to nonferrous, malleable and steel foundries. Several hundred foundries would come under this classification, and when it is remembered that it ordinarily requires the handling of approximately 150 tons of material to produce one ton of gray iron castings at the shipping door, it will be recognized that this material must be handled efficiently in order to produce castings at low cost.

7. In the new order of industry today, with plants operating on shorter working hours per week, materials handling necessarily will play a big part. Since cost of production determines the success of any business, both time and cost considerations will be dealt with in this paper.

Handling Raw Materials from Cars

8. To secure information on equipment for handling raw materials on their receipt at the track, the writer wrote to several equipment manufacturers, endeavoring to find something suitable for the type of foundry being described. In each case, however, he was unable to obtain definite, satisfactory information.

9. At present, the least expensive method of getting raw materials from cars to the storage bins is to bring the cars in on an overhead switch and drop the materials direct from the cars into their respective bins. This is employed by many larger melters, but the expense of erecting the necessary facilities is too great for the average semi-production foundry.

10. Perhaps the next best method of unloading is by crane. Where space permits, this method can be used quite economically for sand, coke, pig iron, etc., especially when the crane can place pig iron, scrap and coke direct on the charging floor. Sufficient labor is saved with this type of equipment for the foundry with smaller tonnages to make substantial savings.

11. Another method of unloading cars of scrap and sand is by the continuous belt, in which the materials are taken directly from the car and dropped in the bin. A simple labor-saving device for handling pig iron consists of the ordinary roller conveyor, allowing the pig iron to roll into separate bins.

12. The question as to which of these devices is most economical for a foundry to use depends upon the tonnage unloaded and the labor rate paid by the foundry.

Charging the Cupola

13. For charging raw materials into the cupola, investigation shows that, at present, hand methods still are used almost entirely. Mechanical cupola-charging equipment, however, has made some very definite forward strides and has been improved considerably since first it was introduced. A number of such systems are on the market that can be used by the average foundry at a saving, although in the author's opinion a cupola would have to melt at least 30 tons per day to warrant the expense of such an installation.

14. In the writer's opinion, there is still much room for improvement of cupola-charging equipment now on the market, and surely there must be some device for economically charging cupolas of smaller tonnages.

15. Perhaps the greatest stride that has been made in the past few years in materials-handling equipment for foundry use is in the distribution of molten metal from the cupola to the molders' floors. Almost every foundry has either a monorail, narrow-gage track, or cranes for relieving the molder from carrying his ladle of iron from the cupola to his floor. The cost of monorails is so low that a foundry can hardly afford to be without them.

16. Relieving the molder of carrying his iron by hand is an improvement that immediately shows up in larger production per day per molder, plus a definite improvement in the finished castings.

Handling Molds

17. The molding department offers what is probably the greatest field for savings through proper materials-handling methods. To make molds most economically, we immediately think of power molding machines for jolting, squeezing, rolling-over and pattern draw. These types of machines are now so flexible that the jobbing foundry with many different size flasks finds these machines very adaptable to its use. The sand-slinger has become quite common in use, the sand-handling feature of this machine being a considerable labor-saving device in itself.

18. Naturally, in using any or all of these labor-saving molding machines, well-made and well-prepared patterns are of prime importance.

19. In connection with the handling of sand by the molder, the old method of using the shovel for filling the mold from a sand heap on the floor was slow and expensive. Certainly, where we strive to get more uniform castings and greater production per day, some kind of an overhead sand-handling system should be employed that will give properly conditioned sand to the molder, when and where and how he wants it.

20. In the past few years some condensed sand-handling systems have been developed that made it possible for foundries employing anywhere from 12 men up, to have sand-handling equipment, if during times of normal operation they can well afford to install them. Some of these systems are quite flexible, and it is possible to start with part of a system and add to it as the times and conditions permit.

Sand Handling

21. Sand-handling equipment is not necessary at all times for continuous pouring. In the case of the continuous pour, the mold must be taken away from the molders' floor by trolley, belt or rollers, conveyed to the pouring platforms and taken over to the shakeout, the sand dropping into the reconditioning hopper and the casting going on to the cleaning room. This type of mold handling is not at all unwieldy, and the upkeep is, as a rule, quite low.

22. In the mechanical handling of sand from shakeout to a molder's floor, new problems arise that must be worked out very carefully, especially as to the conditioned sand to be furnished the molder.

23. Sand must not be so hot that it cannot be used. It must not be balled up so that it makes castings of poor surface appearance. It must not be too wet nor too dry for the particular job required, and, of course, here is where the problem of the jobbing foundry that runs varied types of castings comes in, as to the type of sand to be furnished each floor.

24. In this entire discussion it should be remembered that, due to the wide range of work, most semi-production foundries require at least two different types of sand, either because of finish or of the permeability necessary on chunky or cored work. The

semi-production jobbing foundries also require great flexibility. Many of these foundries sell their entire production as rough iron castings. It may happen, however, that in any six-month period the customer list may change considerably and therefore change the type of product.

25. All these matters must be studied over very carefully before definitely deciding on the type of equipment that is applicable to any one job, in any one foundry.

Summary of Molding Department Requirements

26. To sum up the molding department, a semi-production foundry requires sand conditioning and handling equipment to capably produce at least two different kinds of sand and keep them separate. It needs great flexibility in order that only one-third or one-half of the unit may be run at one time if necessary; otherwise, small systems for 8 or 16 molders should be installed. It needs flexibility so that different types and sizes of patterns can be run. It needs a simple method of removing the molds from the molder and of returning the sand, bottom boards, pouring jackets and flasks to the molder, and a means of removing fine dust at the shakeout.

27. The author sincerely believes that the question of sand handling in the foundry is only in its infancy, and that it will not be very long before a foundry, in order to exist, will have to handle its sand mechanically.

The Shakeout

28. The next operation to be considered is the shakeout. If the shakeout is not a part of a continuous molding system, monorails provide about the simplest method for transporting castings from the shakeout floor to the cleaning department. If included in a continuous molding layout, a system of trucks or rolls or monorails can be used for moving these castings—always remembering, however, to handle each casting, sprue or piece of equipment the fewest possible number of times.

29. Tied in with the system of bringing the castings to the cleaning department should be a system of rolls, monorails or trucks to facilitate handling through the cleaning, grinding, chipping, inspecting and shipping departments. The departments other than molding require a simple system of rolls, monorails or trucks so that nothing is handled twice where the same results can be

accomplished by handling once, and also so that all back tracking is eliminated and straight-line production achieved.

30. In the core room the first operation lies in mixing the core sand; all core rooms should be equipped with one of the various types of core-sand mixers. If the mixing department can be placed in a room above the core room, it makes a very satisfactory arrangement for dropping the core sand into hoppers directly over the coremaker's bench. From the coremakers to the oven, a system of rolls or trucks could be installed to eliminate any long-distance carrying of cores. In addition, where production on individual jobs allows, the core room should be equipped with core-blowing machines or with jolt roll-over machines. The core-blowing machine has proved to be a fine materials-handling device and has shown large savings. Cores should be carried to the foundry on a truck or monorail.

Costs for Continuous Molding

31. The average cost of installing complete equipment for continuous molding has been from \$2000 to \$3000 per molder's floor. At first glance this may seem to be a high equipment cost, compared with the savings possible in molding labor today. However, under the conditions which are likely to prevail in the near future, this price may be low per floor in considering what the labor cost is likely to be.

32. It is also highly important that the foundries should produce more cheaply, so as to compete with other materials which are continually invading the castings field.

33. I believe most foundrymen will agree that much of the upkeep cost with materials-handling equipment has been the fault of the foundries themselves. In many cases the upkeep of this rather elaborate and expensive equipment has been left to rough, handy men who do not exercise the proper amount of care. There is no reason why a materials-handling system, once it is properly installed, cannot be kept going with a minimum upkeep cost if it is properly taken care of from the beginning.

34. In closing, it should be emphasized that the comparison between the initial cost of the equipment, plus upkeep, against the molders' earnings, is not all that can be figured in on a job. Mechanical equipment should provide a considerable percentage in additional pounds of castings per day per molder. It should make possible better castings, and it should allow the foundries to take

care of their customers with less pattern equipment. Savings resulting from the use of mechanical equipment appear not only as dollars saved in labor but in many other ways as well.

ORAL DISCUSSION

MEMBER: How many tons of sand do you use per molder?

R. G. WEILAND: We are using about 8 tons. We operate a jobbing foundry, not a specialty shop.

MEMBER: Do you attempt to keep your production within certain weight limits on castings or do you consider a casting of any weight in going out to get business?

R. G. WEILAND: We have an average, light production shop, running squeezers, floors, benches and roll-over machines. We try to get orders where there will be 50 or more molds on a single job. We generally take castings below the 200 to 300 lb. weight range.

A. F. ANJESKEY:¹ Do you not think that the foundrymen's attitude toward maintenance and wear is due to the fact that in the past, say from 1927 to 1930, they thought too much about continuous equipment, while during the past 3 years, they found that a lot of this equipment had to be operated at 100 per cent although it possibly was producing as low as 5 per cent of its normal capacity yet required the same maintenance? That became a heavy expense. Foundrymen were thinking too much in terms of automobile production as far as castings were concerned in their earlier installations and forgot entirely that they should have more flexibility.

R. G. WEILAND: That is true. Although many installations have been made that did not get a good start and mechanical equipment was ruined before operation was begun.

A. F. ANJESKEY: When our representatives approached a foundryman on simple installations, the founder usually had the idea of continuously moving equipment.

MEMBER: What was the cost per molder for equipment?

R. G. WEILAND: I said \$2000 to \$3000 per molders' floor.

CHAIRMAN J. B. WEBB:² You speak of a molders' floor. How many molders constitute that unit?

¹ Tramrall Division, Cleveland Crane & Engineering Co., Wickliffe, O.

² Jervis B. Webb Co., Detroit.

R. G. WEILAND: A molders' floor would be from 10 to 20 molders and the unit itself would cost \$20,000 to \$25,000. That would be everything in the foundry but would not cover anything in other departments.

MEMBER: It would not include any changes to cupolas, would it?

R. G. WEILAND: No. Those are figures that I received from several reliable sources. This would not include flasks or molding equipment.

MEMBER: We found that figure to run from \$1500 to \$4000 per molder depending on the size and elaborateness of equipment.

MEMBER: Does the author believe mechanical handling equipment to be practical for charging a cupola under 30 tons capacity?

R. G. WEILAND: Cupola charging equipment could not be installed unless daily capacity is 30 tons or more. I feel an attempt should be made to develop equipment for the smaller tonnages.

MEMBER: As assistance to the men on the cupola floor, do you feel that it would be of an improvement to install equipment that would bring the material to the mouth of the cupola?

R. G. WEILAND: I believe most shops have either an elevator, rail or conveyor of some kind, which relieves the cupola men of picking up the material in their hands or with a fork and throwing it in a cupola.

A. F. ANJESKEY: It is my opinion that you could buy cupola charging equipment for the smaller cupolas for \$2000. That includes the tram-rail track and charger. Our greatest difficulty is educating the foundryman to mechanical charging. There may be advantages in hand charging and a lot of foundrymen will not change. I know of one foundry that is doing all charging by hand and they are running well over 30 tons.

MEMBER: There are certain castings on which mechanical charging does not work well. I know of one foundry that pours up to 600 tons per day that does all its charging by hand.

A. F. ANJESKEY: I understand that there is one radiator plant, running on a continuous production basis, that uses mechanical charging, while in another plant, the management is of the opinion that mechanical charging is no good and they charge by hand.

CHAIRMAN WEBB: Wouldn't it be a good idea to understand what charging by hand is? You do not mean that cupola men pick up a shovelful of material and throw it into the cupola, do you? They bring the charges up to the cupola by what method?

MEMBER: They use cranes.

CHAIRMAN WEBB: That would be partly mechanical handling. What you mean is the actual operation of putting the material into the cupola.

A. F. ANJESKEY: That usually is done by placing the charge in a bucket. The bottom of the bucket is arranged so that as soon as the bucket strikes the top of the material that already has been charged, the bucket opens automatically or the operator pulls a string when the bucket is 2 or 3 feet above the last layer in the cupola and the charge drops. Some say that bridging results from mechanical charging, others that the iron is not placed properly, that holes and other conditions develop. I think that it largely is a matter of foundrymen making up their minds that they can use mechanical charging equipment. Some have taken the attitude that it cannot be done.

R. G. WEILAND: It is true that a lot of foundrymen prefer hand charging, although I think that there has been a trend to mechanical equipment recently. With a good bucket, one that gives distribution, they would be glad to put in charging equipment.

MEMBER: The question of cupola charging is one that must be studied carefully. Our cupolas run about 150 tons per day, and most material handling engineers would say that it would be economical to use a charging system. We have made a careful study of our conditions and one of our plants operating on that tonnage has a complete charging system, but we are doubtful whether it pays us much to use the mechanical charger. You must consider the coke ratio. A charging system will do the work, probably, but will burn too much coke. On the other hand, if you use hand charging, you can cut your coke ratios down a whole lot. There are a whole lot of angles to that problem.

As the speaker was saying in regard to the question of using a charger on a 30-ton heat, I doubt whether it would prove satisfactory. For example, on our charging systems, and also I can speak for another large radiator company, they maintain about a 5 to 1 ratio. On hand charging we get up to 12 to 1. It is not a matter of eliminating a few men but a question of coke ratios and overall economies that must be considered. My advice to the small plant would be to study conditions carefully and see what the saving in labor is and how that will balance against other charging methods.

CHAIRMAN WEBB: Why does mechanical charging equipment increase coke consumption?

MEMBER: It is because you cannot place successive layers of coke and iron in the furnace the way you can by hand. It is an impossibility.

A. F. ANJESKEY: What must the bucket do in laying the charge? Must you place the charge uniformly?

MEMBER: In charging the cupola, it is desirable to get uniform layers of coke and iron. Buckets that I have seen cannot place the alternate layers of coke and iron in the cupola with the same uniformity that is possible with a good, well-regulated hand system.

R. G. WEILAND: In hand charging, the fatigue element is present particularly in smaller plants where there is a lack of concentrated supervision and where the cupola men many times charge the furnace as they please.

CHAIRMAN WEBB: The author spoke of high maintenance in foundries on mechanical material handling equipment. My experience in material handling has been similar to Mr. Anjeskey's. I had occasion to recommend some equipment to a foundry and the man in charge was an experienced foundryman with a reputation for low maintenance costs. It was my desire to save the firm money and I recommended a weight and type of equipment which I felt was husky enough from the standpoint of pure engineering. The foundryman said, "Foundry practice is different entirely from machine shop or general bulk handling because of the abrasive qualities, the chances of wrecks, etc." To my surprise, he decided to buy equipment about three times as heavy as that recommended. The result is that his maintenance has been extremely low.

I think that a little more cooperation on the part of the foundry in-

dustry, the material handling firms and a closer analysis of the actual problems involved would benefit everyone concerned. Materials can be carried with certain types of equipment readily under normal conditions but the foundry situation is abnormal, from a mechanical standpoint. The foundry requires stiffer and stronger equipment that will withstand a tremendous amount of abrasion. I liken it to the cement industry, where everything must be built to withstand high abrasion.

MEMBER: While price buying was an evil of foundry mechanization, I think most difficulties may be laid to maintenance. I can remember two identical systems of about the same weight. One went to pieces in a short time and the other is still running. It all goes back to maintenance. I checked back and found one fellow greased the belt conveyor idlers once a week whereas the other fellow didn't touch them for six months because he happened to see a folder that said they could go without greasing for six months.

CHAIRMAN WEBB: That is a very good point. In a foundry, there is not the tendency to care for equipment because it is built for abuse. If the material handling industry will take that into consideration, it will help the situation.

Cupola Refractories

By EDWARD E. MARBAKER,* PITTSBURGH.

1. The average foundryman is accustomed to consider the problems incident to the lining of cupolas as more or less a matter of course. He is not inclined to give such problems the attention they deserve from the cost standpoint. More attention should be given to the selection of refractories from the standpoint of service rather than from that of price alone.

2. Some years ago a sub-committee under the Joint Committee on Foundry Refractories was appointed by the AMERICAN FOUNDRYMEN'S ASSOCIATION to investigate the field of cupola refractories. A great mass of data was collected and studied but no direct conclusions could be drawn from it. It was found that the cost of refractories per ton of castings varies from \$0.05 to \$1.50, and the cost of daubing varies from \$0.06 to \$1.68, averaging \$0.20 per ton for refractories and \$0.40 per ton for daubing. Taking these figures at their face value, the wide variation from the average shown by the highest and lowest values indicates that the operator of the first foundry might well give his refractory cost a great deal of attention, and that the second probably did not have his cost figured correctly. The figures also show that the cost of daubing or relining is about twice that of the original refractories, a situation that would be expected because the lining of the average cupola is replaced perhaps once a year, while its melting zone is relined after every heat.

3. The cost of refractories is related closely to correct cupola operation. If a cupola is operated so that the melting zone bridges over early in the heat, not only the melting efficiency suffers, but the cost of relining the melting zone is increased materially. Moreover, even the highest grade of refractory material will be worth

* Vice-president, Industrial Research and Engineering Co.

NOTE: This paper was presented at the session on Foundry Refractories at the 1933 Convention of the American Foundrymen's Association.

little to the foundryman if his cupola is not operated in accordance with the best practice. When the cupola is properly lined, maintained, and operated, real economy will result.

4. The lining of the cupola above the melting zone is not subjected to very high temperatures, or to slag action, if the limestone flux is added properly. Its main service is to act as an insulator and a support, and it must withstand the abrasion of the continuously moving charge. For these reasons, the brick or block need not be of a highly refractory character, but must have a high crushing strength and resistance to impact. It also must be fairly dense.

REQUIREMENTS OF CUPOLA REFRACTORIES

5. For that section of the cupola immediately below the charging door, many foundrymen have found it advantageous to install several courses of cast iron blocks which withstand the impact and abrasion of the incoming charge. This is especially true where certain types of charging machines are used.

6. In the melting zone, the lining is subjected to high temperatures, 2900-3100 degrees Fahrenheit. For that reason, the use of a highly refractory material is vital, particularly when long heats are taken off. Slag action in this section is most violent, and the lining must be dense and well laid with as few and as narrow joints as possible. The lining also must be able to withstand spalling caused by the extreme changes in temperature above the tuyeres and by the rapid cooling to which it is subjected after the cupola bottom is dropped. A much higher grade of brick or block is required for service in this part of the cupola than for the part between the top of the melting zone and the charging door.

7. The melting zone should be lined with first quality clay refractories to withstand the high temperature. To counteract slag action, a fine-grind brick or block would be indicated, but the tendency toward spalling would be increased. A compromise is required and generally a medium-grind refractory of high softening point has been found to give the best all-round service.

8. Attempts to introduce high-priced special refractory materials such as silica or silicon carbide have met with scant interest because of the cost. They might prove to be highly economical but few foundrymen are willing to experiment with them. Manufacturers are not equipped generally to carry out experimentation that would show conclusive results.

9. Cupolas may be lined with standard firebrick or with cupola block. The latter are made to provide walls of the desired thickness either in one or two vertical courses. Often the lining or the shell is built of standard brick and the remainder of cupola block. The advantage of cupola block lies mainly in the production of relatively few joints. That results in diminished slag penetration and a lower cost of erection.

TAP-OUT AND SLAG-HOLE BLOCK

10. It has been customary for years to tap out through a breast which was made by the cupola tender when he prepared the cupola for the day's heat. This is a job that requires considerable skill. Recently, it has been found more convenient to use refractory blocks of proper size and provided with one or two tap-holes. This type of block can be installed as a part of the lining and does not have to be replaced after each heat. For this service a refractory of high grade must be used, because in addition to stability at high temperature, it must possess resistance to abrasion of molten metal and chemical action of the slag.

11. Similarly a slag-hole block is used in the opposite side of the cupola wall to permit the removal of slag. This also is installed as an integral part of the cupola lining. It must be capable of resisting slag action and mechanical abrasion. The latter is especially important in operations in which the slag is drawn off intermittently and the slag hole must be opened from time to time by an iron bar.

12. In continuously operated cupolas, it is customary to use a slagging spout. In this case, the iron and slag run simultaneously from the cupola and are separated by a damming arrangement in the spout. The iron passes under the dam and the slag is conducted through an opening in the side of the spout above the dam. This device usually is lined with standard firebrick which are protected by daubing. The latter is replaced after each heat. A high grade of material should be used in this service.

DAUBING

13. Maintenance of the lining of the melting zone costs roughly twice as much as the actual lining of the cupola. The subject is worthy of considerable attention, and, unfortunately, it seems to receive less than it deserves.

14. The lining of the melting zone is subjected to the hardest possible service. High temperature, slag penetration, abrasion and

spalling are all factors upon which its service life depends. After each heat it is found that part of the lining has disappeared and it has to be rebuilt.

15. Daubing too often consists of a poor grade of fire clay, obtained locally at a low price, mixed with a grog of broken-up firebrick or silica rock, and water to make a thick paste. The use of good material would constitute an economy there.

16. High quality fire clay is very refractory and, when it is used in daubing, it will not tend to soften at high temperature and become a prey to abrasion and slag action. This will result in the need for smaller daily applications. Moreover, the protection of the costly lining of the cupola is much more effective.

CUPOLA OPERATION AND REFRACTORY SERVICE

17. Correct cupola operation has a definite bearing on refractory life of the melting zone. Improper combustion, faulty slagging conditions, or a low-fusing daubing tend toward the formation of a bridge. This upsets the balance of the cupola all through the heat. The bridge generally gets worse, the longer the heat. Finally, it may close the cupola completely. Bridging is accentuated by the use of low fusing fire clay in the daubing.

18. Proper operation of the cupola necessitates the use of correctly-designed refractory materials which will withstand the severe service to which they are exposed.

ORAL DISCUSSION

J. M. MCKINLEY¹: In looking through the TRANSACTIONS of the A.F.A., A.S.M.E., American Ceramic Society, British Ceramic Society, Institute of British Foundrymen, and the French and German institutions, I was startled to discover that there is practically no fundamental technical data dealing with the reactions upon refractories in a cupola. The entire cupola reaction is vague. Variations in practice are extremely wide. I have only to mention that one of our customers manufactures car wheels and uses an extremely high grade brick. Another equally important customer, also a car wheel manufacturer, uses a brick at the other end of the scale.

In the 1928 issue of the TRANSACTIONS of the A.F.A., there is an article* by C. E. Bales in which he defines four groupings as representing the divisions under which specifications might arise for cupola refractories: (1) Intense heat; (2) chemical abrasions; (3) thermal shock; and (4) physical abrasions.

Throughout the paper and in other literature dealing with the subject, there seems to be the general supposition that the properties underlying

¹ North American Refractories Co., Cleveland.

* *Refractories for the Cupola*, by C. E. Bales, TRANSACTIONS, American Foundrymen's Association, vol. 36, 1928, pp. 683-696.

such divisions can be considered separately. Generally speaking, research work in refractories during the past 5 to 7 years would dispute that conclusion. Those properties, with a number of others, are interlocked so closely that it is impossible to consider any one of them alone.

It generally is conceded that the cupola operates at about 3100 degrees Fahr. I have not been able to find any data pertaining to surface heats developed on the refractories. Manifestly, if the burden temperature is close to 3100 degrees Fahr., the same temperature will not exist at the outside surface of the lining as will exist at the face. A contrast is found in open-hearth work where surface heats generally are conceded to be about 2850 degrees Fahr.

As a matter of fact, research work has not progressed far enough in that direction to assure us that that figure is accurate. We do not know authoritatively that silica and practically all fire clay brick will not withstand over 3100 degrees Fahr. surface heat. Foundry cupola practice represents one of the most severe applications for refractories.

Improper practice will lower the surface heats but, at the same time, it generates gas conditions that will have to be considered before any definite progress can be made. It is known authoritatively, from the standpoint of refractory research, that ferrous silicates developed from a combination of any silica present and coke ash have extremely low melting points. Many cokes have an ash melting point as low as 2280 degrees Fahr. and a few in excess of 2550 degrees Fahr. It is not difficult to conceive that an extremely active material will be generated under such circumstances.

I would like to call attention, for some future thought, to the fact that limestone is known to decompose at about 1550 degrees Fahr. In ordinary metallurgical work, little attention has been paid to what happens when that release takes place. Based on research work over the past 3 years, largely having in mind blast furnace lining material, I would like to call attention to two definite lines of thought.

First, CO_2 does not remain constant. If conditions are perfect, it probably will pass through without reaction. But conditions generally are not constant and at all times in the cupola, there is a certain degree of CO generation. Carbon-monoxide actively disintegrates the lining. Penetration of the CO into the brick occurs and an internal explosive force is set up. I have found many cases where cupola blocks were supposed to have spalled. The bricks showed all the evidence of spalling but it was not thermo-spalling. Carbon-monoxide penetration was the cause. That caused a preliminary disruption of the material which lead to a rapid physical break-up as the burden was lowered and especially when the cupola was quenched.

On the lime side of the problem, it is known definitely that there is an active absorption on the part of all clay refractories and a consequent breakdown later. Coupled with the action of both carbon-monoxide and lime, a disputed element enters the picture. Work that has been done seems to indicate that cyanogen, $(\text{CN})_2$, is developed in the cupola.

Positive evidence has been found of the existence of potassium cyanide. Both of those elements, especially the $(\text{CN})_2$, are aggressively

active in the carbon-monoxide penetration. The acceleration is tremendous. That theory alone, if it can be reduced to a plant practice definition, may solve a great many problems of cupola refractories.

J. T. MACKENZIE:² In cupolas operating at 1590 degrees Cent. (2900 degrees Fahr.) where the iron is tapped the condition probably is entirely different from one tapped at 1400 degrees Cent. (2550 degrees Fahr.). We find that at 1375 degrees Cent. (2500 degrees Fahr.) the molten iron undergoes a great change. Above that temperature, oxidation of the air causes burning of carbon. Below that temperature, silicon and manganese are oxidized. It would make a lot of difference in the cupola lining depending on whether carbon or silicon and manganese are oxidized. I do not know that those temperatures would hold in the cupola. With probably 15 per cent carbon-monoxide and 12 or 15 per cent carbon-monoxide, those reactions might take place at 1600 degrees Cent. (2900 degrees Fahr.). Nobody seems to have studied the subject. The foundryman who is getting fine results from one kind of refractory may be oxidizing only carbon in the cupola. Another foundryman who has experienced ruinous results and has changed to another type refractory, may be oxidizing silicon and manganese.

It might be possible to have refraction at the temperature at which the slag is being thrown up the wall. The wind catches the forming slag and throws it up. If the refractory is sufficiently porous that the slag gradually melts and forms a tough, sticky, viscous, impermeable layer, that particular refractory might give excellent service. However, if the slag simply sticks to the brick and acts entirely locally, the first piece of pig iron or coke that strikes that brittle spot will knock it off. Then the opportunity for critical reaction seems to be almost infinite. I do not believe that those conditions will ever be divorced from refractories.

Car wheel foundries are content to run on 1350 or 1400 degree Cent. (2460 or 2550 degree Fahr.) iron. They make a big casting. They draw the mold, open the whole bottom of the ladle and pour the car wheel all at one time. They have efficient cupola operation around those temperatures. I think that car wheel foundries use less coke per ton of iron than most foundries. That means higher carbon dioxide, more iron oxide, etc.

Take the foundry that is making automobile castings. A certain foundry melts iron at 1585 degrees Cent. (2885 degrees Fahr.). It would be safe to say that the gas ran 20 per cent carbon monoxide. Probably no iron was oxidized, only carbon.

CHAIRMAN A. V. LEUN:³ Carbon deposition does not occur to any great extent at temperatures over 1200 or 1300 degrees Fahr. or under 800 degrees Fahr. That means a certain distance back from the face of the lining, most of the carbon deposition will occur, with subsequent depositing. In addition, at least 20 per cent CO must be present in the gas to retard deposition.

J. M. McKINLEY: Carbon-monoxide is not necessary if you have $(CN)_2$ present. A half of one per cent $(CN)_2$ will cause more damage than 20 per cent carbon-monoxide.

² American Cast Iron Pipe Co., Birmingham, Ala.

³ Bethlehem Steel Co., Bethlehem, Pa.

Factors Affecting the Service of Clay Refractories in Iron Foundries

By R. C. ZEHM,* WELLSVILLE, O.

1. Refractories of the type used in iron foundries are generally made from two materials; fireclay, a raw material, and grog, which may be considered as a semi-finished material.
2. For some purposes, a refractory made only from fireclay, without the addition of grog, may be desirable. That is true particularly in the development of a product which is resistant highly to slag action and penetration.

CLAY PROPERTIES VARY

3. The type of fireclay for refractories adaptable to the variable service conditions existing in iron foundries, is dependent to a great extent upon the chemical analysis and physical properties which are transmitted to and developed in the finished product. The use of a fireclay of the desired type is conditional upon uniformity.

4. There are two types of grog: calcined clay and burned brick or fireclay shapes. Calcined clay, when ground and screened, generally produces grog particles which, by virtue of their sharp edges and angular shape, are best adapted to the proper bonding of clay and grog and the manufacture of a desirable product.

* The Stevenson Co.

Note: This paper was presented at the session on Foundry Refractories at the 1933 Convention of the American Foundrymen's Association.

5. Practically no relationship exists between crushing and the quality of the finished product.

6. Grinding and screening invariably constitute a closed cycle in which the improper functioning of either phase nullifies the possibility of obtaining proper results in the other phase.

7. The size of the clay and grog particles, their shape and the proportions of the various sizes of the particles, have a direct bearing upon the finished product, with reference to resistance to slag penetration and absorption, the ability to withstand mechanical shock, abrasive resistance and other conditions of service.

GRAIN SIZE IS IMPORTANT

8. Particles of clay or grog which are too large, or an excess of larger sized particles, are conducive to the development of a product which has a high resistance to rapid temperature changes but they tend to an open structure of the product which reduces mechanical strength and increases slagging action and abrasion. Improper and non-uniform proportioning of the various sized clay and grog particles are contributing factors to inferior quality.

9. The desired proportions of clay and grog required to develop a suitable refractory product generally are determined by volume measurement. Mixing and grinding tend toward a closer blending of clay and grog and has an influence on the quality of the final product.

THREE PROCESSES AVAILABLE

10. The manufacturing process is continued by one of three methods. All processes involve the addition of a definite and uniform quantity of water to the mix.

11. The three methods are known in the clay industry as the (1) hand made, (2) the stiff mud and (3) dry press processes. The type of product desired and service conditions to which it is subjected are the determining factors in selecting the process. Each of these processes imparts to the finished product improvements in quality of product which are not readily obtained by either of the other processes, and in some cases not obtained at all. Conversely, each one of the three processes influences unfavorably certain properties of the finished product as compared to the same properties which are developed in the refractory by the other methods of processing or forming.

12. After the brick or shapes are dried properly, they are

placed in the kilns and burned. The time-temperature curve for the firing of clay refractories is planned carefully and controlled accurately to produce a high percentage of No. 1 ware, which has chemical and physical properties permitting it to be efficiently and economically used in the refractory service for which it was developed.

13. The upper part of the cupola requires a refractory which is highly resistant to mechanical shock and abrasion. A suitable refractory for the upper section of the cupola should be as uniform as possible throughout and should not contain inherent defects which are developed when the refractory is subjected to the temperature and other service conditions existing in that part of the cupola.

SELECTION OF REFRactories

14. For the cupola melting zone, a refractory for satisfactory service must withstand the temperature developed and also resist effectively slag action caused by contact of the slag and the refractory and also caused by the reaction with the refractory of the secondary slag which is produced by the initial action of the melt slag upon the refractory.

15. The area of contact of slag with the refractory varies with the mechanical structure and density of the refractory. If the grains of a refractory are not thoroughly embedded in and surrounded by the clay matrix, slag will penetrate into the refractory, increase the area of contact and accelerate the slag action, thereby decreasing the length of service of the refractory. The temperature range existing between the vitrification point and the melting point of a refractory is often a factor in the ability of a refractory to withstand slag action. In general, a refractory with a great difference in temperature between its vitrification and melting points is less susceptible to slag action.

16. The installation of refractories often decreases their length of service, especially when they are handled roughly causing breakage of the corners and edges. Serviceability also is decreased if much mortar or a mortar inferior to the brick or blocks is used in laying up refractories. The joints between the brick or blocks should be of minimum width.

ORAL DISCUSSION

CHAIRMAN A. V. LEUN:¹ In one particular case, where brick was installed in a malleable furnace, we were using hand-made brick and getting approximately 18 heats per slag line. We changed to a stiff-mud brick with 6 to 7 per cent more density and we doubled the life. That brings out the point that, to resist slag erosion, a fairly dense brick is required. A brick of that type is paramount to resist spalling.

MEMBER: One of the principal factors affecting the life of bung brick in malleable furnaces, is the spring in the arch. We made a survey in 1927 and found the furnaces with the longest life were those that had the greatest arch. From my observation, the failure of bung brick does not result from refractory quality as much as it does from abuse, crane handling, etc.

G. OLSON:² What can brick makers give the malleable founder in the line of side walls that will withstand corrosive slag? We have found a great many times that, by changing from one type brick to another, we have entirely different causes of failures in our side walls. Not so long ago, we bought some brick and I was able to increase the life of the side walls by just patching the slag line and leaving the remainder of the wall untouched.

MEMBER: Unless a high-priced brick gives considerably better service than a low-priced brick, the high priced material would not be economical. It is my opinion that the labor would be approximately the same. In most cases, labor costs of patching are approximately equivalent to the cost of the material used. If considerably better service is not secured with good brick, probably the difference in material used would not offset the difference in price.

G. OLSON: Referring to bricks for repairing the cupola, those brick are not placed there without joining material. It does not matter whether good or bad brick are used unless a mortar is utilized that can be depended upon to be at least as good as the brick. It is not the brick that causes the trouble as much as it is the mortar.

MEMBER: If I was going to use good brick, I would use fire-clay brick. Sometimes good brick and good clay are used in a cupola and it is hardly possible to tell the difference in erosion from where second grade brick is used.

J. T. MACKENZIE³: We had an interesting experience with a carload of overburned brick. We had been using from 600 to 700 standard brick per day in our cupolas for patching. We found after operating 3 weeks on the overburned brick that we were using only 450 per day. The brick were almost vitrified.

CHAIRMAN A. V. LEUN: In the steel mills, brick that had been burned and vitrified hard and was kiln marked, would not be used in ladles. A plant at Bethlehem used kiln marked brick and increased the ladle lining

¹ Bethlehem Steel Co., Bethlehem, Pa.

² Beloit Foundry Co., Beloit, Wis.

³ American Cast Iron Pipe Co., Birmingham, Ala.

life by three or four heats. That is definite proof that the denser the brick, the lower the porosity and the slower the erosion.

MEMBER: One man may run a cupola for 6 hours and tap 10 tons of metal and another may operate for 14 hours and tap 120 tons. The amount of metal tapped makes a great deal of difference. Those two installations are not comparable as far as cost is concerned. The whole problem depends upon how much labor can be saved by putting in a high priced lining that will last longer.

Advantages of Radiographic Inspection to Foundrymen

By HERBERT R. ISENBURGER,* NEW YORK.

Abstract

Exposure and cost charts for the examination of steel castings give a comparison of the relative economics between x-ray equipment and radium. A practical example of gamma-ray field inspection is shown. X-ray exposure technique and foundry practice are discussed and demonstrated in some cases. A new photographic paper is introduced for the first time. Its possibilities on steel and non-ferrous castings are shown. The value of x-ray inspection as a production tool and as a sales factor is mentioned.

1. In radiographic inspection, two kinds of radiation are dealt with, x-rays and gamma-rays. The important difference, which must be considered for radiographic inspection, is wave length of the rays. X-rays excited by a voltage of 120 kv. have a wave length about 0.0001 that of ordinary light and 1200 kv. would be required to obtain x-rays which have the same wave-length average as gamma-rays. Wave-length usually is measured in Angstrom units. One Angstrom unit equals 0.00000001 cm., or 10^{-8} cm. Fig. 1 shows a curve of x-ray wave lengths at various voltages. This demonstrates that the higher the voltage the less the change in wave length. In the low-voltage range, the difference in wave length is so great that a slight difference in voltage makes a noticeable difference in the resulting x-ray picture. X-rays excited by 30 kv. have a wave length of 0.41 Angstrom units, for example.

2. The longer the wave length the better the definition in the negative. This would mean that for heavy material, a long

* St. John X-ray Service Corp.

NOTE: This paper was presented at one of the sessions on steel castings at the 1933 Convention of the American Foundrymen's Association.

NOTE: The superimposed numbers throughout this article are references to the bibliography which appears at the conclusion of this paper.

exposure time rather than a short exposure with high voltage should be used. This is possible only in a limited sense. Looking at the exposure charts in Fig. 2, with a certain voltage, a satisfactory picture through $\frac{7}{8}$ inches of steel in one minute is obtained, whereas another voltage will penetrate $1\frac{1}{8}$ inches in the

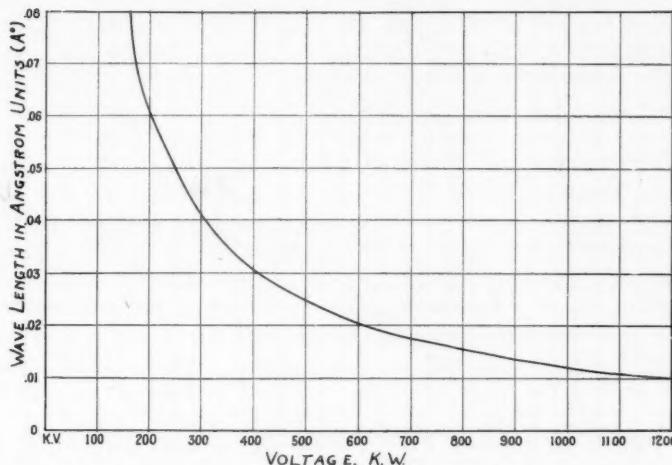


FIG. 1—CHART SHOWING X-RAY WAVE LENGTHS AT VARIOUS VOLTAGES.

same time and in another, $3\frac{1}{4}$ inches. Any material that is thinner than stated, should be x-rayed with a 1-minute exposure using a lower voltage rather than the same top voltage and a shorter exposure time. The necessary voltage for other distances and exposures can be determined from equations but is read more conveniently from charts such as Fig. 2.

3. For routine inspection work, a 1-minute exposure is preferred. A 10-minute exposure is considered a desirable limit from the economic point of view. With 120 kv. at 5 ma., this limit would be $1\frac{1}{8}$ inches; with 200 kv. at 6 ma., $2\frac{3}{8}$ inches; and with 300 kv. at 10 ma., about 4 inches. In certain cases, it may be permissible to go beyond these limits even for routine inspection. In Fig. 2, the curve for radium inspection is shown on the same scale as the x-ray curves but instead of minutes, the ordinates are hours. X-ray as well as radium curves are given for exposures at 24 inches focus-film distance. Values for other distances can be calculated by applying the inverse square law.

4. In the following text, some practical applications of radiographic examination of steel castings are shown. In one case, an 8000-lb. cross fitting was installed in a boiler house which was under construction. The casting was hanging about 40 feet above the floor and it was practically impossible to install x-ray equip-

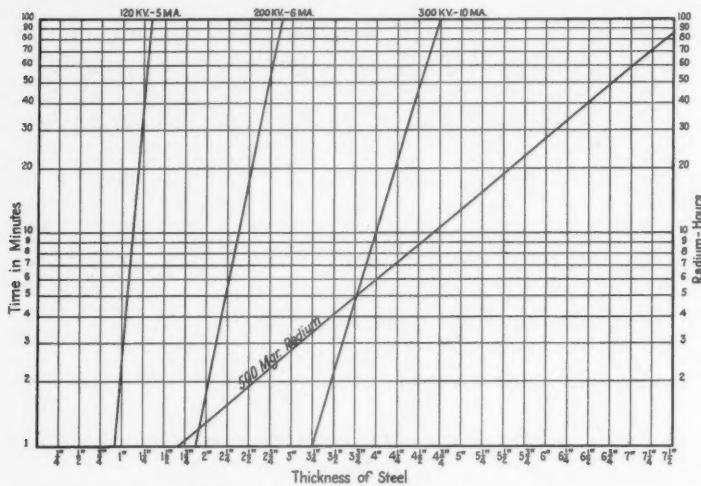


FIG. 2—CHART SHOWING EXPOSURE TIMES FOR VARIOUS THICKNESSES OF STEEL. 24-INCH FOCUS-FILM DISTANCE; NORMAL FILM DENSITY, 0.7; INTENSIFYING SCREENS USED.

ment for its inspection, although the wall thickness was within the limits of x-ray routine. Radium had to be used.

5. Identification lead markers were fastened with adhesive tape on the surface of the steel casting and the fitting was covered completely with film holders. Various amounts of radium, suitable for over-night exposure, were placed at five different locations inside the casting and the whole fitting was examined in one exposure. Although the inspection disclosed some serious defects, quite a number of fine but objectionable cracks were missed and were discovered by subsequent physical tests.⁵

6. To obtain sharply defined outlines of the shadows when using x-rays, lead sheets are employed to screen off any stray or secondary radiation which might otherwise fog the film and cover up details. That is not necessary with gamma-rays. A typical exposure arrangement for the inspection of the high-pressure seat of a safety valve body is shown in Fig. 3.⁶ Fig. 5 shows the

exograph through the cope side of this casting when set up as shown in the diagram. The lead identification marker 1.3 was placed on the body between the neck and the flange.

7. The white streaks indicate shrinkage cracks or primary pipe due to improper feeding. Thirteen of these castings were sent to our laboratory for x-ray examination. All showed the same serious condition in identically the same place. The mold was redesigned and the riser head made larger so that the gas could escape more readily. All replacements were perfect. Had some pilot castings been inspected before production was started, considerable time, machining and other expense would have been saved. Similar shrinkage cavities usually can be avoided by casting with the heavier sections uppermost and by providing large feeding heads over the heavy sections.

8. In another case, seven out of eleven, 6-inch, T-fittings for high pressure steam service showed duplicate defects, indicated by an arrow in Fig. 6. The defect is illustrated in exograph Fig. 7. It is a gas pocket with shrinkage cracks extending into the flange. This condition subsequently was overcome by putting more metal into the fillet.

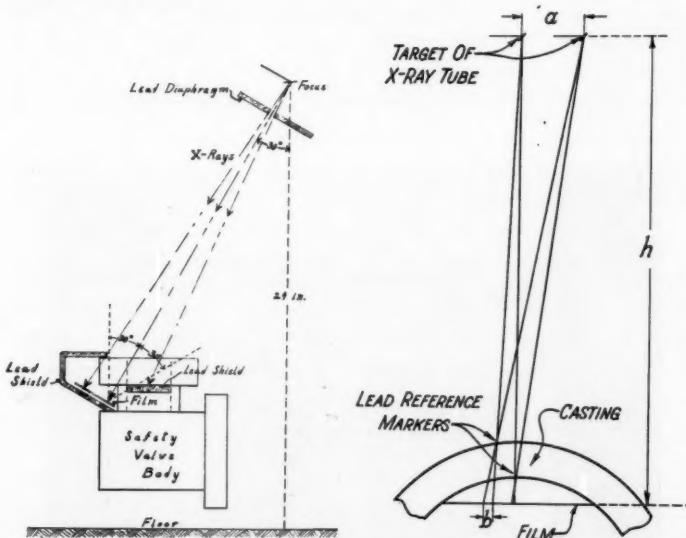


FIG. 3—LEFT—DIAGRAM SHOWING EXPOSURE CONDITIONS FOR X-RAY INSPECTION OF A SAFETY VALVE BODY. FIG. 4—RIGHT—SET-UP USED FOR DOUBLE-EXPOSURE X-RAY METHOD TO SHOW DEPTH OF DEFECT.



FIG. 5—EXOGRAPH TAKEN THROUGH RISER SIDE OF HIGH PRESSURE SEAT OF VALVE BODY.

9. An exograph of good steel⁸ is shown in Fig. 8. The parallel bands are screw threads cut on the inner surface of the valve body. The light spot in the lower left hand corner is not a defect. It is the inner end of a hole drilled through the wall.

10. Sometimes a casting with holes or irregular shapes of varying thickness must be x-rayed. The holes, etc., are filled with a material of approximately the same atomic weight as the object to be examined before making the exposure. A simple method of doing this has been developed by Dr. Aneel St. John.⁹ He immerses the specimen in a liquid absorber. In the case of steel, he



FIG. 6—SIX-INCH TEE-FITTING ARRANGED FOR X-RAY INSPECTION.

uses a lead salt solution containing 3 lbs. of lead acetate and 3 lbs. of lead nitrate in 1 gal of water.

11. Fig. 9 shows a part of a tractor shoe. The casting weighs about 30 lbs. and consists mostly of webs which are up to 2 ins. thick. To obtain sharp outlines of the webs, the whole casting was immersed in a solution which absorbed the transversing rays. That was necessary as otherwise the x-rays would have been scattered by the webs and the entire picture would have been fogged.



FIG. 7—EXOGRAPH TAKEN THROUGH 1 1/4 INCHES OF STEEL IN REGION 3-1
INDICATED BY THE ARROW IN FIG. 6.

A few gas inclusions are visible in some of the webs. In the case of gamma-rays, that precaution would not be necessary.

12. Many imperfections disclosed by radiographs are permissible. The boiler code permits a reasonable degree of porosity or discontinuity in fusion welds. Agreement on allowable conditions for castings would be desirable. For instance, a cavity or slag inclusion may be permissible if not too deeply seated. For this purpose, a double-exposure x-ray method has been developed.¹⁰ Through the use of angulation, it enables the inspector to determine the location as well as the depth of the defects. The method is simple and involves only the making of two exposures of the same area on one film with a shift of the x-ray tube between exposures.

13. The set-up for this method is illustrated schematically in Fig. 4. During the exposures, the tube is moved parallel to the film at a predetermined distance a . The distance between the target and the film also is known. The film then records two images of the same irregularity in the material under investigation. The distance between the two images on the film can be measured

and will be found to be b inches. The distance to be determined is the space between the cavity and the film. This unknown distance is designated as x . The following proportion then exists:

$$a : b = (h - x) : x \text{ or}$$

$$x = \frac{hb}{a + b}.$$

Since the distance b usually can be neglected in comparison with a , the equation becomes

$$x = \frac{h}{a} b = kb \text{ approximately.}$$

14. A typical example is shown in Fig. 10. The subject is a section of a 3-ton, steel casting with a $2\frac{1}{2}$ -inch wall thickness, containing a fairly large sand pocket. The shadows of a reference lead marker were displaced $\frac{9}{16}$ inch or more and the spot



FIG. 8—EXOGRAPH TAKEN THROUGH $2\frac{1}{2}$ INCHES OF SOUND STEEL.

below the marker, $\frac{3}{8}$ inch. This spot was on the inner surface. Various spots in the sand inclusion were measured and their displacement was found to be close to that of the outside marker. Thus it was determined that the inclusion started close to the outer surface and did not penetrate more than one-quarter of the wall thickness. This left sufficient sound metal between the sand

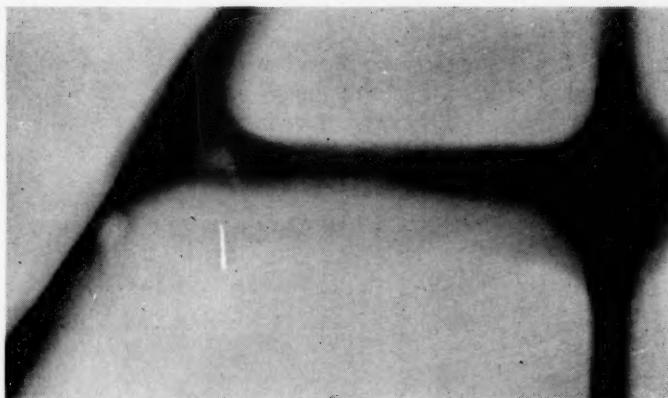


FIG. 9—X-RAY PHOTOGRAPH OF A SECTION OF A TRACTOR SHOE IMMERSSED IN A LIQUID ABSORBER.

pocket and the inner surface. The defect was not considered to be of a serious nature.

15. Not only can steel foundry technique be controlled and bettered by radiographic examination, but gray iron, malleable iron and nonferrous foundry practices can be checked and improved.

16. The builders of a new subway system brought two pieces of broken sewer pipe to our laboratory. The pieces were gray iron, $1\frac{3}{8}$ inches thick and had been taken from a 36-inch pipe line which was built in 1886. The city claimed that the water main broke because of the construction work going on nearby and sued the builder for damages.

17. X-ray evidence, Fig. 11, clearly indicates that the material failed through a weakness inherent in a part of the section examined. Both pieces showed a region of spongy metal adjacent to the breakage crack. The sponginess consisted of a number of gas and sand inclusions with fine shrinkage cracks. There also were some other cracks present within this area. It is evident that most of those conditions were present at the time the material was installed. Some of the cracks may have developed while the casting was in service. Undoubtedly, this region must have been corroded through quite a distance when the breakage occurred and the pipe would have failed through this section without any exterior cause.

18. It is out of the question to examine pipe, which probably costs \$0.02 per lb, when x-ray inspection may amount to half that

cost. However, it pays to x-ray pilot castings, as has been recommended in other instances.¹¹

19. The question if radiographic examination cost is not easy to answer, since it depends on a number of factors that are variable in almost every case. Our company has attempted to establish reliable cost charts per single exposure at various thicknesses



FIG. 10—TYPICAL EXAMPLE SHOWING DEPTH OF DEFECT AS OBTAINED BY DOUBLE-EXPOSURE X-RAY METHOD.



FIG. 11—X RAY PHOTOGRAPH SHOWING SPONGY METAL ADJACENT TO BREAKAGE CRACK IN CAST IRON PIPE CASTINGS. THE WHITE PORTION AT THE CENTER OF THE ILLUSTRATION IS NOT A DEFECT BUT SPACE BETWEEN TWO BROKEN PIECES OF PIPE.

of steel. The appropriate exposure time readily can be calculated for various conditions from the charts supplied in Fig. 2. Consider three different classes of x-ray equipment and radium as follows:

- Class 1: To operate at 5 milliamperes and 120 kilovolts peak.
- Class 2: To operate at 6 milliamperes and 200 kilovolts peak.
- Class 3: To operate at 10 milliamperes and 300 kilovolts peak.
- Class 4: Using $\frac{1}{2}$ gram radium:

20. Tables 1, 2 and 3 give some idea of radiographic inspection cost. The complete amortization is divided over a period of 5 years, not counting the interest on investment. X-ray equipment may be good for another 10 years and radium will last practically indefinitely and always will represent a considerable value. Insurance, electric current and possible repairs are expressed in upkeep. The insurance rate for radium is higher than for the x-ray equipment. Salaries provide for an operator and a helper. The hourly figures are minimum prices for full time operation at 200 hours per month. They are exclusive of handling and labor overhead. If the x-ray equipment is not used continuously, the upkeep cost will be less. If the operating crew is not kept busy with inspection, they can do other work in the meantime, thus reducing the salary figure. If the equipment is in permanent use day and night, working for more than 200 hours per month, the operating cost will be considerably reduced. This is particularly important in the case of radium where long continuous exposures are common.

21. A comparison between Tables 2 and 3 shows how much

cheaper the operating cost is when x-ray equipment and particularly radium has been written off. Charts in Fig. 12 are based on Table 2. They do not include handling and labor overhead which varies with each job and is different in most foundries. They also do not include the cost for film or paper.* The cost

* A sensitive x-ray paper recently has been developed which compares favorably with celluloid film.^{4,12}

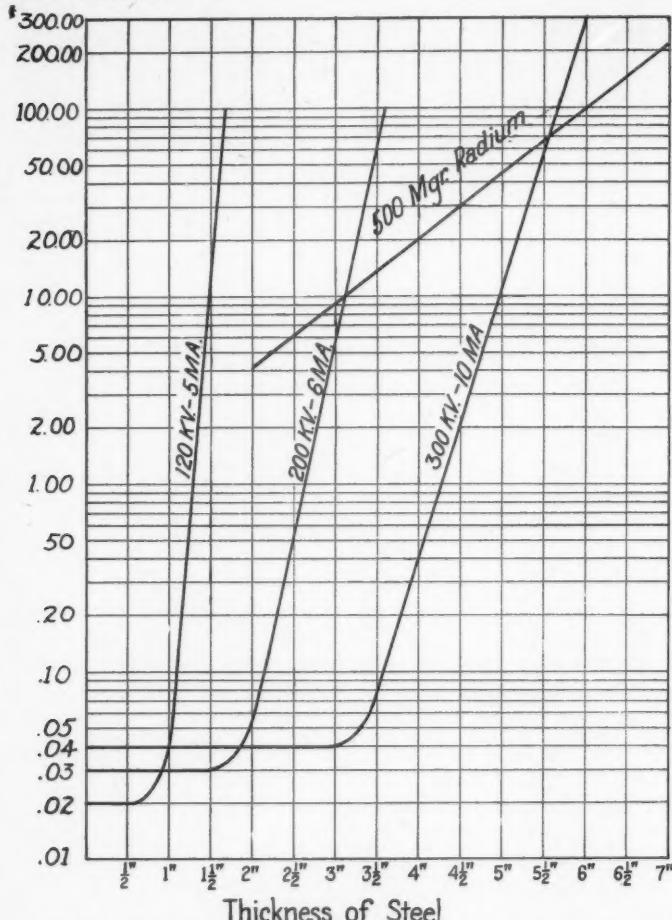


FIG. 12—CHART SHOWING RADIOPHASIC COSTS PER EXPOSURE FOR STEEL CASTINGS, FILM AND HANDLING CHARGES EXTRA. 24-INCH FOCUS-FILM DISTANCE.

for film is about $\frac{1}{3}$ cent per sq. in. and for paper about $\frac{1}{6}$ cent per sq. in.

22. Since radiographic inspection gives valuable information to the foundryman without destroying the casting, and often saves time and money, it should be a worthwhile production tool for the foundryman. In addition to this advantage, the fact that the foundry can offer radiographed castings bears a tremendous sales argument which will break down undue resistance in the mind of doubtful customers. The rapid development of high-pressure, high-temperature steam installations, high-head hydraulic power,

Table 1
COST OF RADIOGRAPHIC EQUIPMENT

Class	Ma.	Kv.	Equipment	Tube	Local Const.	Lead-house	Darkroom	Total Cost
1	5	120	\$1,000	\$150	\$ 250	\$ 600	\$1,000	\$3,000
2	6	200	3,000	260	740	1,000	1,000	6,000
3	10	300	5,000	500	1,500	2,000	1,000	10,000
4	$\frac{1}{2}$ gr. radium	35,000	1,000	36,000

Table 2
OPERATING COST DURING THE FIRST FIVE YEARS

Class	Amortization	Upkeep	Salaries	Total Yearly	Monthly	Hourly	Per 1-min. Exposure
1	\$ 600	\$ 500	\$1,780	\$2,880	\$240	\$1.20	\$0.02
2	1,200	800	2,320	4,320	360	1.80	0.03
3	2,000	1,300	2,460	5,760	480	2.40	0.04
4	7,200	1,800	2,520	11,520	960	4.80	0.08

Table 3
OPERATING COST AFTER THE FIFTH YEAR

Class	Upkeep	Salaries	Total per Year	Monthly	Hourly
1	\$ 620	\$1,780	\$2,400	\$200	\$1.00
2	980	2,320	3,300	275	1.38
3	1,560	2,460	4,020	335	1.68
4	1,920	2,520	4,440	367	1.84

high-pressure oil cracking systems, high-power locomotives, aerial transportation and various other lines of industrial activity, present increasingly numerous cases where reliable castings will be required. Willingness to produce x-ray evidence of quality will be a useful aid in the sale of castings.

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WRITTEN DISCUSSION

R. A. GEZELIUS¹ and C. W. BRIGGS¹: Mr. Isenburger states, in paragraph 2, "The longer the wave length the better the definition in the negative," thereby implying that the definition obtained with X-rays is considerably better than that obtained with gamma-rays. We can agree with neither the statement nor the implication.

The *definition* obtained in any radiograph is dependent primarily upon the diameter of the focal spot of the tube used, or the diameter of the radium pellet, and the distance from the source to the film. The wave length does not affect the definition except that the longer wave lengths may cause blurring due to scattering and secondary radiation. As this is the case, the definition obtained with either method should be approximately the same.

The latest data available at the Bureau of Standards gives the following diameters for the focal spots of tubes of varying wave lengths:

140 kv. tubes can be obtained with a fine, medium or broad focal spot of 7.1, 9.1 and 11.9 mm. diameter, respectively.

¹ U. S. Naval Research Laboratory, Anacostia, D. C.

200 kv. tubes can be obtained with 11.9 mm. diameter focal spots.

300 kv. tubes can be obtained with 13 mm. diameter focal spots.

There is one exception to the above and that is a foreign made tube known as the "Metallex" which can be obtained with a 4 mm. diameter focal spot. This tube, according to reliable information, has not proved to be satisfactory.

The radium used by the Navy is in small silver capsules the largest of which has a diameter of 8 mm., the size of the radium source being but 6 mm. in diameter. This is smaller than any focal spot obtainable in x-rays except with long wave lengths or the "Metallex" tube. Radium and radon (radium emanation) can both be obtained in spheres of even smaller diameter. If long needles of radon, such as are used extensively by the medical profession, are used for radiographic purposes only, the ends of the needles should be used as sources of gamma rays. Any other use of the needles cannot be considered as anything but poor technic.

The *contrast* obtained in a radiograph is dependent entirely upon the wave length used and becomes greater as the wave length increases. Mr. Isenburger may have meant *contrast* rather than *definition* in his statement.

In Fig. 2, Mr. Isenburger shows a series of curves giving the exposure times required for x-rays and gamma-rays at a film-to-source distance of 24 inches. This distance is unusually long for gamma-ray radiography as the distances most frequently used in actual practice are from 12 to 18 inches. Therefore, the exposure times most frequently used are, using the inverse square law, from 25 per cent to 56 per cent that of the times shown on the curve. This statement should not be construed to mean that gamma-rays are not used at distances greater than 18 inches. In one case, where the entire side of a stern post was to be radiographed in position with one exposure, the film-to-source distance was 5 feet.

We note with surprise that in using gamma rays for inspecting a large casting that "quite a number of fine but objectionable surface cracks were missed." The Naval Research Laboratory, as pioneers in gamma-ray radiography, has found that all defects which can be considered objectionable will be found to appear on the films. It might be pointed out that in using several sources of gamma rays on the same casting, as was done in this case, the best of results cannot be expected unless precautions are taken to screen the rays so that each film receives rays from one source only. If this is not done, blurred images and excessive fogging must be expected.

As no explanatory data are included, we are at a loss to understand the figures Mr. Isenburger presents as a comparison of the actual costs of the two methods. Even so, there are several things that we should like to point out.

The initial cost might be a little high as the Naval supply of 0.48 grams, delivered, cost considerably less than the figure given.

We cannot understand why the amortization of radium should be listed as \$7,200 per year for the first 5 years when the actual average depreciation, calculated on its half-life of 1580 years and using Mr. Isenburger's figure of \$35,000, is \$11.08 per year. We do not believe that any

concern would attempt to amortize a \$35,000 investment in radium in 5 years when at the end of that period the radium on hand would be worth \$34,945.

The upkeep on the radium purchased by the Navy has been zero. The only possible upkeep would be insurance and this is deemed unnecessary for the following reasons:

1. Radium for radiographic purposes is in such a form that it is unsuitable for almost any other purpose. It would be certified at the Bureau of Standards and resale, which usually invokes certification at that bureau, would be apt to lead to identification. In case of theft, the exact description of the capsules could be broadcast quickly to all radium users.

2. It is dangerous to handle in any manner except that outlined for its use, and extremely dangerous and probably fatal to tamper with the capsules themselves.

3. Manipulation of the radium capsules in an attempt to make identification difficult could so mark a man that he could be picked out of thousands, unmistakably, with scientific instruments.

4. To find a buyer for stolen radium would be most difficult and to find a buyer who would remain unknown to the American Medical Association and Government agents would be practically impossible.

5. In case of loss, if the approximate place of loss is known, it can be traced and found by scientific instruments.

We cannot understand why the salaries listed for gamma-ray work are higher than those listed for any other x-ray service since gamma-ray technic is learned easily and does not require a trained technician. At the Norfolk Navy Yard, which has a portion of the Naval radium supply and where gamma-ray radiography is more or less routine, an apprentice molder does all of the actual radiographing.

Therefore, using Mr. Isenburger's figure for salaries, we find that gamma-ray radiography should not cost more per year than \$11.08 for depreciation and \$2,520 for salaries, or \$0.016 per minute.

WRITTEN DISCUSSION BY THE AUTHOR

H. R. ISENBURGER: In reply to Messrs. Briggs' and Gezelius' discussion, the second paragraph of my paper should have read contrast instead of definition.

With regard to the size of the focal point in x-ray tubes and gamma ray sources, I wish to refer to an ideal construction of an x-ray tube focus which was described by the author 4 years ago in an article entitled, "X-rays and X-ray Inspection in Piping," in the magazine *Heating, Piping and Air Conditioning*, Vol. 1, 1929, pp. 115 to 120. That type focus now is being used in the modern Coolidge x-ray tubes. Thus the size of the focal spot of a 300 kv. tube is smaller than the gamma ray source used by the Navy, taking advantage of the so-called line focus principle.

The size of the focal spot of a gamma ray source is only ideal when the radioactive material is especially packed for industrial radiography. In most cases, we have to utilize radium capsules and needles which

are ordinarily used for medical purposes, that is, for treatments and not for photography. The necessary amount for a particular job can be rented for a certain time. Depending on the strength of the source required, the focal spot may be $\frac{1}{2}$ -inch in diameter and 1 or $1\frac{1}{2}$ -inches long. This broad focus accounts for not being able to detect fine defects such as were discovered by subsequent visual examination in the cross fitting mentioned in the paper referred to previously. Radon is too expensive in most cases, since this material must be purchased outright.

As to the focus-film distance, all our x-ray standards have been based on a 24-inch distance. For comparison, we had to choose the same distance for a gamma ray chart. The same reduction of time is effective in accordance with the inverse square law for a shorter focus-film distance when using x-rays as in the case of gamma rays. Dr. Mehl, in his curve, uses the following distances for gamma ray work: 12, 18, 24, 30 and 36 inches and 24 inches is the happy medium.

Concerning the cost for radium, the discussers are correct. The price is much less today than the author stated. In 1930, when scouting around for radium, the price given us was \$70,000 per gram. We were told that this price would be maintained by the Belgian radium monopoly. We are informed that today's market price is \$50,000 per gram. If we had bought the radium and followed the discussers write-off procedure, today we would have lost about \$20,000. In addition, America went off the gold standard, that is, the market price as compared with the international value for radium is another 30 per cent less so that the actual value of one gram of radium today is exactly one-half its value 3 years ago.

I wish to correct my tables on the bases of \$25,000 for $\frac{1}{2}$ gram of radium.

Table 2
Operating Cost During the First Five Years.

Amortization	Upkeep	Salaries	Total per Year	Monthly	Hourly	Per Minute Exposure
\$5200	1800	2360	9360	780	3.90	0.06 $\frac{1}{2}$

Table 3
Operating Cost After the First Five Years.

.....	1920	2360	4280	357	1.78
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If a commercial foundry would buy radium, they would never write off this material over a period of 1580 years, as suggested by the discussers. One obvious reason has been shown. Others are the fact that even today's price of \$50,000 is merely artificial established and dictated by the radium monopoly. Should new sources be made available in this country, in Canada or in Russia, this artificial price could not be kept. Furthermore, new and more economic ways of obtaining radium or other radioactive materials may have a considerable influence on the price fluctuations. It is more likely that a foundry purchasing radium will write it off against a specific job for which they need it than over a period of 5 years. It certainly would be better business practice.

No business man can afford not to insure his radium. A physician who possesses a little over a half a gram of radium told the author that his material cost him \$20 daily for upkeep alone, whether he uses it or not. Since this figure may be exaggerated, we feel that our figures given for upkeep are fair.

In case the radium supply is stolen, this material readily can be packed in a number of small containers. When sold at half the market price, they will find any number of buyers. There is a lucrative market for second hand radium now in existence. The Bureau of Standards would naturally certify any medical container, for instance, which would be presented for that purpose.

Figures for the salaries include labor. When using radium, we presume that heavier material is to be examined which requires more help for handling than does material ordinarily examined with x-rays. Therefore, the slight increase of \$60 over a period of an entire year. This figure, however, has been adjusted to suit the final figure in the corrected tables.

The Experience of a Gray Iron Jobbing Foundry with the Rocking-Type Electric Furnace

By C. R. CULLING,* ST. LOUIS

Abstract

This paper shows practical applications of the rocking-type electric furnace for production of a wide variety of castings to meet the requirements encountered in the general jobbing field. The first major point brought out is the possibility of manufacturing a product having greatly improved qualities. This improvement is brought about through increased strength, increased density as desired, and improved machinability in the denser, stronger materials. Some factors involved in production of this improved quality are the available superheat, deoxidizing atmospheric condition, and absence of slag and oxide inclusions. The next salient point is the establishment of definite control over quality. Characteristics of this furnace are such that the meeting of definite specifications and duplications of both chemical and physical properties are assured after determination of the proper mixture. The third point emphasized is the flexibility of the rocking electric furnace, as evidenced by the wide range of chemical analyses covered in production of the regular grades of higher strength materials, including various alloys. In addition it was found both practical and economical to produce small batches of metal for special purposes or to produce a larger quantity of a base mixture and progressively change the bath to suit different requirements. The last thought is one of concern in this time of curtailed production, namely, the possibility of increased service to consumers. By use of the furnace described, it has been found possible to serve customers consistently without waiting for larger accumulations of tonnage necessary for operation of the cupolas.

* Superintendent, Carondelet Foundry Co.

NOTE: This paper was presented at one of the sessions on Cast Iron at the 1933 Convention of the American Foundrymen's Association.

1. During recent years there has been a determined effort on the part of the alert gray iron foundryman to overcome the prejudices existing against the use of gray iron for anything other than ordinary requirements. These prejudices have existed, for the most part, for a great many years, and were probably justified by a lack of interest and knowledge on the part of the foundryman.

2. It was not deemed possible to obtain gray iron of high strength and which was not brittle. Cast iron was considered only as a cheap material, easily produced to conform to patterns and suitable for use chiefly where strength and shock resistance were not required. Handbooks and tables used by designing engineers gave low strength values for this material; hence, parts made of it usually were massive to provide desirable factors of safety.

3. The effort to overcome these prejudices led naturally to the development of equipment and technique for the manufacture of a new product, entirely divorced from the old conception of gray cast iron.

Quality Control of Cast Iron Needed

4. In order not only to keep up with the march of progress but also to keep ahead of it, we realized some three years ago the necessity of producing iron castings of entirely different characteristics than had previously been made. We saw a distinct advantage for the foundryman who could produce iron castings with higher tensile strength and greatly increased density with good machinability. We further realized the absolute necessity of having highly accurate control over the composition and physical properties of the metal and, hence, over the ultimate quality of the castings.

5. With these requisites under consideration, we thoroughly studied the subject and undertook development work which led to the installation of an electric furnace for the production of the type of cast iron we desired to produce.

Capacity of Electric Furnace Installation

6. Our electric furnace installation, illustrated in Fig. 1, is of the indirect-arc rocking-type, manufactured by the Detroit Electric Furnace Co., and is rated at 3000 lbs. capacity. In it we melt charges ranging from 500 or 600 lbs. to 4000 lbs. Our usual charge is 2000 lbs., which we melt in approximately 1 hour and 20 minutes. The nature of the materials charged varies with the

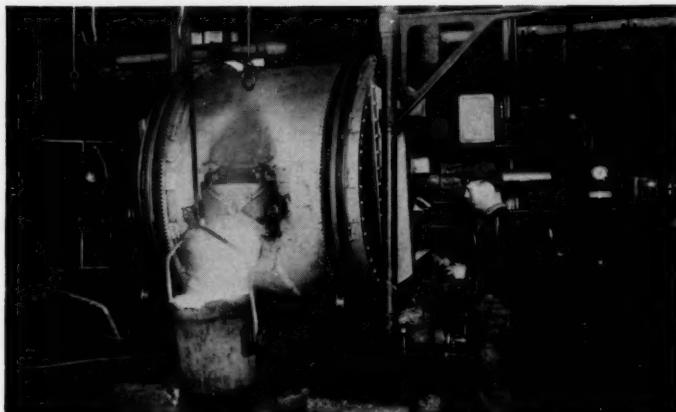


FIG. 1—ROCKING ELECTRIC FURNACE, 3000 TO 6000 LBS. CAPACITY, POURING INTO 700-LB. BULL LADLE SUSPENDED FROM MONORAIL. NOTE SLATE ABOVE OPERATOR'S HEAD, USED BY HIM TO INDICATE THE COMPOSITIONS WHICH ARE TO BE POURED FROM THIS HEAT.

requirements of the metal produced, but consists chiefly of cast iron borings, steel scrap, cast iron scrap, and sprue, as well as the additional alloys used in the manufacture of special castings.

Carbon Control

7. It is well known that the characteristics of cast iron change chiefly through manipulation of the carbon in the iron. This involves not only varying the amount of carbon contained but also controlling the state in which the carbon exists in the castings. This is done by superheating the iron, deoxidizing it, controlling the rate of cooling, and by the use of silicon or other graphitizing agents.

8. Having available the means of superheating and deoxidizing our iron, the method of procedure that suggested itself was to determine, for each type of casting, the correct carbon content and then to establish the correct silicon content to suit the mass and section of the casting and the rate of cooling. In some instances we find it advantageous to employ iron containing carbon as low as 2.00 per cent, while in other cases it is as high as 3.50 per cent. The silicon varies from 0.75 per cent to 3.00 per cent.

9. A casting of $\frac{1}{4}$ -inch section may require 3.00 per cent carbon and 2.00 per cent silicon to produce maximum strength with good machinability, whereas a casting of 2-inch section may

require 2.50 per cent carbon with 1.50 per cent silicon in order that maximum properties may be realized. It is interesting to note that in the production of castings of approximately $\frac{1}{4}$ -inch section we have observed that our electric furnace iron, containing 3.10 per cent carbon and 2.00 per cent silicon, can be cast in certain designs with excellent machinability and showing no chilled edges. Ordinary iron of the same composition, poured into the same castings, is white and unmachinable. This is a most interesting phenomenon and is probably best explained by the superheat and deoxidizing action of the electric furnace.

Duplication of Results

10. Having established the correct composition for any casting, we found it desirable to set up a procedure that would make its reproduction at any future time a matter of routine. We began by assigning numbers to a series of irons of a range of compositions suitable for general application. As need arose, many of these were modified to suit new types of castings, and new numbers were assigned to these modifications. We have established some seventy-five standard compositions containing carbon from 2.00 per cent to 3.50 per cent, with varying silicon, nickel, chromium, copper and molybdenum contents.

11. As stated, each composition is numbered and jobs having special requirements are indexed according to the metal number established for the job. The pattern clerk, melter, chemist and foundry foreman each has access to one of these indices, and when the job goes to the foundry the shop ticket is marked for the proper iron. The foundry foreman collects his data as to the required amount of each metal and then gives orders to the melter to produce the necessary quantity. The melter records the iron number on his daily report sheet, and as each heat is charged into the furnace he shows its number, in large chalk figures, on a slate which hangs beside the furnace.

12. When ready, the iron is distributed to the proper floors and molds under the supervision of the foreman. The chemist checks the iron from each heat and is aided by knowing in advance what alloys to determine.

13. An average day's production of ten tons will have possibly six base mixtures, with three to five additional nickel-chromium and molybdenum mixtures. Record of the base heats

Table 1
BASE HEATS MADE DURING A TYPICAL DAY'S PRODUCTION

Mixture No.	TC	Si	Ni	Cr	Mo	Use.
8	3.25	2.00	1.50	Cams and gears.
10	3.50	2.60	Thin castings.
21	2.80	1.75	Pressure goods, medium.
22	3.00	1.75	Pressure goods, light.
52	2.60	1.50	1.50	0.60	Dies.
68	3.00	1.75	0.50	Machinery frames.
33	3.10	2.00	Misc. medium-sized castings.

of a typical day's production is given in Table 1. It will be noted that irons numbered 21 and 22 are very similar, varying only in carbon content. Both are used for pressure goods, but each is designed to produce equal density, strength and machinability in castings of considerable difference in average section.

14. Whenever a pattern comes in for additional orders it is routed through the shop according to the mixture number. Thus, the customer is assured that the castings produced will have the same composition, physical properties and machinability as those previously furnished. Our ability to meet these specifications is assured because of the characteristics of the rocking indirect-arc furnace in which we melt the iron.

15. We find that this furnace is a precision tool with which we can achieve any desired predetermined analysis and which can be duplicated at will. For example, one type of metal is made by melting 100 per cent steel punchings and adding carbon with the charge by calculation. The carbon recovered is 76 per cent of the coke added. A typical charge consists of 2000 lbs. of steel punchings and 58 lbs. of crushed coke. This produces a metal containing 2.20 per cent total carbon. Silicon as required is obtained from ferrosilicon added on the basis of complete recovery.

16. We depend entirely upon calculation for the production of this iron, and our results from more than 200 heats of it show that the carbon of the iron is within the limits of 2.18 and 2.24 per cent.

Ranges of Application

17. As was previously stated, we have standardized some seventy-five different types of iron, ranging from 2.00 per cent to 3.50 per cent carbon and from 0.75 per cent to 3.00 per cent silicon,

with numerous alloy combinations. The possibilities for application of these precise specifications of composition to a wide range of work are obvious. This flexibility of melting control makes possible the long-standing and oft-expressed desire of the foundryman operating a jobbing shop, namely, the ability to select and produce the type of metal best suited to meet the needs of the various classes of work encountered in the jobbing shop.

Small Quantities of Various Irons in One Heat

18. It has proven practical and economical to produce small quantities of the different metals by progressively changing the bath during the heat. For example, we may have a 4000-lb. heat containing 2.75 per cent carbon and 1.50 per cent silicon. After tapping 1500 lbs., 10 lbs. of crushed coke are added, the power turned on, and the furnace rocked for three or four minutes. This produces a bath containing 3.00 per cent carbon, the silicon remaining constant at 1.50 per cent. Another 1000 lbs. may be tapped and 15 lbs. of 50 per cent ferrosilicon added. After rocking the furnace for three or four minutes, the metal will be uniformly of 2.00 per cent silicon, with the carbon remaining constant at 3.00 per cent.

19. In this manipulation, the carbon and silicon are accurately and effectively increased as desired. Both the carbon and silicon are readily assimilated by the bath because of the mixing action of the rocking furnace and because the surface of the bath is clean and free from slag.

20. In other instances, depending upon shop conditions, we may have a 3000-lb. bath containing 3.00 per cent carbon and 2.00 per cent silicon. After tapping 2000 lbs., we may add 115 lbs. of

Table 2
ILLUSTRATING ACCURATE DUPLICATION OF A GIVEN COMPOSITION
AT INTERVALS DURING TWO MONTHS

Heat No.	TC	Si	Mo
881	3.00	1.76	0.54
889	3.01	1.78	0.55
896	2.97	1.80	0.55
902	3.01	1.78	0.54
910	3.00	1.80	0.55
919	3.02	1.78	0.56
1047	2.98	1.78	0.50
1056	3.03	1.76	0.56

steel punchings, turn the power on, and rock the furnace two or three minutes to produce a bath containing 2.70 per cent carbon and 1.80 per cent silicon.

21. Near the conclusion of many of our base heats the remainder of the bath may have alloys added, such as nickel, chromium or molybdenum. It is well known that much of the benefit of such alloys is lost when ladle additions are used, because of incomplete solution and chilling the iron. The full value of all alloys is obtained in our electric furnace because of its mixing action and because the temperature of the iron is not lowered.

22. These examples of our routine practice have been given to show that our control over the composition of the iron has been reduced to a matter of mathematical precision.

Production of an Alloy Iron

23. A compilation of eight 1-ton heats of an alloy iron, produced over a period of about two months, is shown in Table 2. In these heats the carbon, silicon and molybdenum were calculated for 3.00, 1.75 and 0.50 per cent, respectively. All of these heats contained approximately 40 per cent sprues and gates of the desired composition, 30 per cent steel scrap and 30 per cent iron borings.

24. This iron is used for heavy-duty punch press frames. When made from ordinary semi-steel, frequent failures resulted and the manufacturer had decided to use steel castings. We convinced him that we could produce an electric furnace iron of suitable strength, and subsequent tests in service have shown these castings to possess strength much in excess of the requirements.

25. The tests for these castings were conducted as follows: The semi-steel castings which were failing were tested by punching boiler-plate material of varying thickness until the gage of material resulting in their failure was established. This material was then formed on the machine with the new electric furnace casting substituted, and the machine was undamaged. The material being punched was then increased in thickness until the manufacturer determined that the pressure exerted was twice as great as that which caused failure of the semi-steel castings, with no failure on the part of the new castings.

26. The manufacturer calculated that a pressure in excess of 100 tons was being applied and, needless to say, was satisfied that the electric furnace high-test iron possessed strength assuring an

ample margin of safety. Fig. 2 illustrates the structure of this iron, which possesses in excess of 55,000 lbs. per sq. in. ultimate tensile strength.

27. Especially in the production of alloy irons do we realize the advantages of the neutral atmosphere and the stirring action of our electric furnace. These characteristics enable us to obtain complete recovery of all alloys added, whether as ferroalloys or in foundry returns, and to produce with absolute uniformity an alloy heat of any desired size.

Some Special Irons

28. For example, in producing such alloys as Ni-Resist, our percentages of nickel, copper and chromium remain constant after

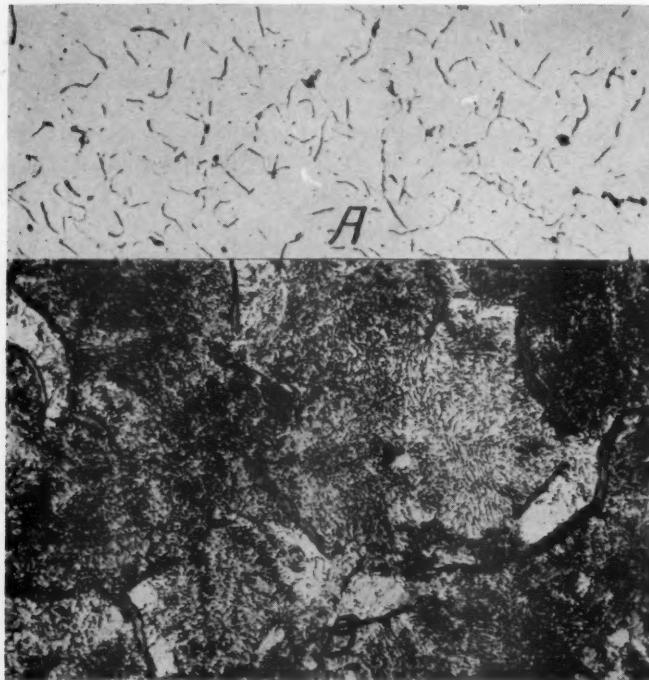


FIG. 2—HIGH-STRENGTH IRON (CARBON 3.00 PER CENT, SILICON 1.75, MOLYBDENUM 0.50) USED FOR HEAVY-DUTY MACHINERY FRAMES. A: SHOWING VERY FINE, EVENLY DISTRIBUTED GRAPHITE. UNETCHED, X100. B: SHOWING WELL-DEFINED SORBO-PEARLITIC STRUCTURE. ETCHED IN 4% NITAL, X500.

repeated remelting of sprues and returns. We also melt high alloy irons and steels for high-temperature and corrosion-resistant applications. Some of them contain up to 35 per cent chromium with or without nickel, while others contain up to 35 per cent nickel with chromium as high as 16 per cent. We find, in the production of these high alloys, that we obtain the same close control of composition previously described, and that we obtain complete recovery of the alloys contained in the return gates and sprues.

Production of Ammonia Valve Castings

29. One of the most logical applications of electric furnace iron is in the production of pressure castings. The requirements of this iron are strength, density, freedom from inclusions and good

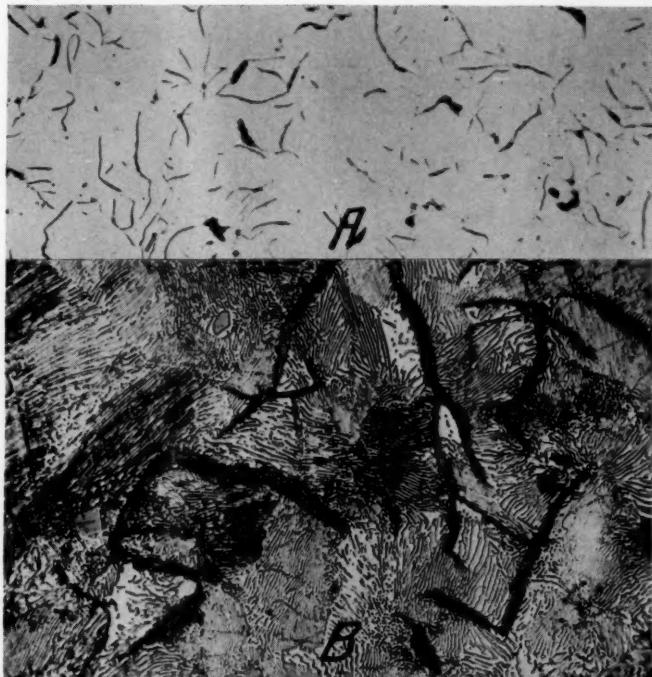


FIG. 3—UNALLOYED IRON (CARBON 2.70 PER CENT, SILICON 1.70) USED EXTENSIVELY FOR HIGH-PRESSURE SERVICE. A: SHOWING FINE, UNIFORM GRAPHITE DISTRIBUTION. UNETCHED, X100. B: SHOWING CHARACTERISTIC NORMAL PEARLITIC STRUCTURE, GIVING EXCELLENT MACHINABILITY. ETCHED IN 4% NITAL, X500.

machinability. We have developed a group of electric irons especially designed for this service.

30. It has been stated that the carbon and silicon content is exceedingly important in producing the desired structure in pressure castings. Fig. 3 shows the structure of a specimen removed from a high-pressure ammonia valve. Note the well-developed structure of lamellar pearlite and unusually fine, thoroughly distributed graphite, both of which are essential. Physical tests of this iron made on A.S.T.M. 1.20-in. bars broken on 18-in. centers show results as follows:

Tensile Strength, above	50,000 lbs. per sq. in.
Transverse Strength	3700 lbs.
Deflection	0.30 in.

31. Fig. 4 shows the uniform hardness of castings made from this iron. As shown, there is a variation of only five points in the series of Brinell readings made on this casting.

32. Some interesting technique has been developed in the production of these ammonia valve castings, where light and heavy

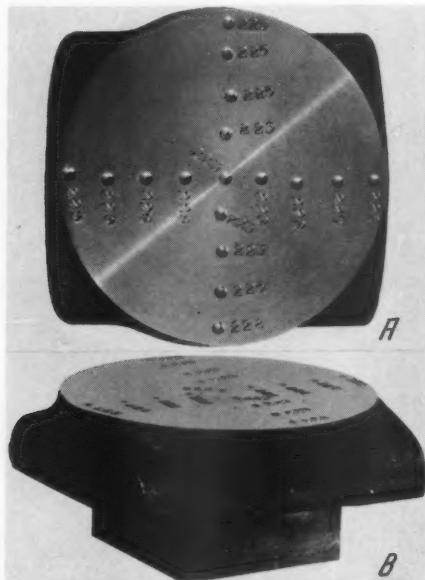


FIG. 4—SAME IRON AS SHOWN IN FIG. 3. A: SHOWS REMARKABLE UNIFORMITY OF HARDNESS, WITH BRINELL HARDNESS VARYING ONLY 5 POINTS IN 17 IMPRESSIONS. B: SHOWS DESIGN AND PROPORTIONS OF CASTING.

sections adjoin. In these cases the iron is designed for the heavy sections and nickel is used to prevent the light sections from becoming too tough and slow to machine.

33. The records of the manufacturer of these high-pressure ammonia valves show that during 1932 the entire loss, due to foundry defects and machine shop spoilage, was less than 5 per cent. Prior to using electric furnace iron, the loss was approximately 25 per cent. In one particularly troublesome casting the losses formerly ran above 50 per cent. This has been reduced to less than half of one per cent.

Die Production

34. Another established use of electric furnace iron is in the production of dies, many of which we supply to manufacturers who were formerly using steel. Some of these dies are used for forming hot metals. A special nickel-chromium alloy iron is used which, with a heat treatment that has been developed, produces dies that retain their hardness under the conditions imposed by this service.

Abrasion-Resistant Castings

35. We supply a great many castings which are subjected to abrasive service or wear, and find this to be a very fertile field. Our application for this in ordinary service is a very low silicon iron with varying carbon content and containing alloys in some instances. Where the requirements are extremely high, we find that a very low silicon with high carbon, combined with suitable alloys, gives the best service.

36. The ordinary unchilled iron without alloys has a Brinell hardness ranging from 400 to 450. By adding nickel and chromium, the Brinell hardness of the unchilled iron is in excess of 600. These materials are used for ore and stone conveyors, bottom plates in coke chutes, pulverizer hammers and many other similar applications.

Sleeves for Large Piston Rings

37. Another very satisfactory application for electric furnace iron is found in sleeve or pipe stock for large piston rings, bull rings, packing rings, etc. We have designed a series of irons for this purpose and have obtained startling results. When made from ordinary iron it is by no means uncommon to see half of a

sleeve discarded because of spongy iron. The sleeves that we supply are uniform in structure from end to end and their entire length is used.

38. For these sleeves we use a low-carbon iron with the silicon regulated to produce good machinability in the section involved. The smaller castings, approximately $\frac{1}{4}$ in. thick and from 4 to 6 inches in diameter, require iron with 3.00 to 3.10 per cent carbon and 2.00 per cent silicon. These combinations are varied with larger sleeves; and one 30 inches in diameter and $1\frac{1}{4}$ inches thick requires iron with 2.70 per cent carbon and 1.50 per cent silicon.

Table 3
CHEMICAL AND PHYSICAL PROPERTIES OF TYPICAL HIGH-STRENGTH IRONS

Mixture No.	TC	Si	Ni	Cr	Mo	Trans- verse,* Strength, lbs.	Deflec- tion,* inches	Tensile Strength, lbs. per sq. in.
33	3.10	2.00	2765	0.312	40,600
67	3.00	1.75	3260	0.290	51,200
68	3.02	1.74	0.67	3760	0.310	57,750
27	2.79	1.75	3365	0.286	52,210
23	2.78	1.75	0.70	3745	0.302	59,905
78	3.20	1.60	0.55	3450	0.309	53,880
52	2.60	1.50	2.00	0.60	4245	0.349	52,960
59	2.75	1.75	1.50	0.60	3450	0.290	53,750

* Tests were made on A.S.T.M. 1.20-in. bar, broken on 18-in. centers. Tensile specimens were machined from the broken arbitration bars.

Strengths of Different Irons

39. The foregoing paragraphs have revolved largely about a discussion of the development of the desired structure for numerous castings of varying sections and for specific applications. Not much has been said about the strength of the different irons.

40. The development of the proper structure in any given casting application automatically results in adequate strength. We feel, as do many other foundrymen, that the use of test bars for measuring the properties of a casting is both hazardous and unreliable, and we, therefore, have definitely established our procedure to produce in the castings themselves the properties desired. In many cases these properties are determined by destructive tests on the castings. However, we occasionally are required to submit test bars with the castings delivered.

41. Table 3 shows the properties of some of these bars made from several of our typical high strength irons, plain and alloyed. Figs. 5, 6 and 7 show the structure of some of these irons mentioned.

Heat-Treated Irons

42. A type of iron produced in our daily routine and which has received wide application is a series especially designed for heat treatment. This metal is made from steel punchings with carbon additions, as previously described, and contains from 2.00 to 2.50 per cent of carbon and 1.00 to 1.50 per cent silicon. The heat treatment is regulated to produce varying degrees of strength, hardness and ductility, and is completed within 24 hours.

43. The characteristics of this iron can be varied within



FIG. 5—A 2.41 PER CENT CARBON, 2.26 SILICON, 0.50 MOLYBDENUM IRON; SPECIMEN FROM CENTER OF 2-INCH DIB CASTING. A: SHOWING EXCEPTIONALLY FINE GRAPHITE AND CELLULAR DISTRIBUTION. UNETCHED, X100. B: SHOWING SORBITIC PEARLITE MATRIX. ETCHED IN 4% NITAL, X500.

wide limits, according to the composition and heat treatment used. One class is made by a routine that produces elastic limit of 40,000 to 45,000 lbs. per sq. in., with ultimate tensile strength of 55,000 to 60,000 lbs. per sq. in. with 12 per cent elongation in 2 inches. Another class has an elastic limit of 50,000 lbs. per sq. in., with ultimate tensile strength in excess of 75,000 lbs. per sq. in. and 5 per cent elongation in 2 inches. A third class has from 3 to 4 per cent elongation, with 65,000 lbs. per sq. in. elastic limit and 90,000 to 100,000 lbs. per sq. in. ultimate tensile strength. Table 4 illustrates typical physical properties of irons of each group.

44. In addition to these irons, all of which show some ductility, we make other heat-treated castings. In many instances we

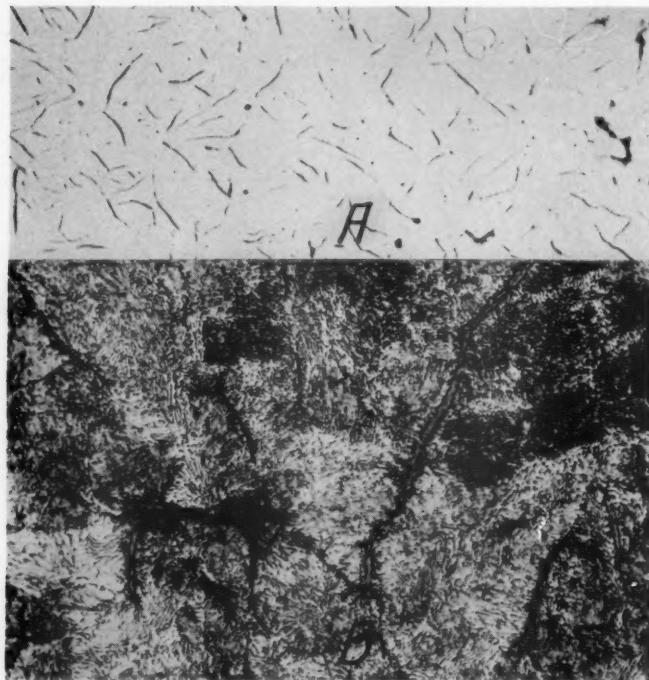


FIG. 6—UNALLOYED GENERAL UTILITY IRON (CARBON 3.20 PER CENT, SILICON 1.54), RUNNING OVER 50,000 LBS. PER SQ. IN. TENSILE STRENGTH. A: SHOWING VERY UNIFORM DISTRIBUTION OF FINE GRAPHITE. UNETCHED, X100. B: SHOWING SORBITIC PEARLITE STRUCTURE. ETCHED IN 4% NITAL, X500.

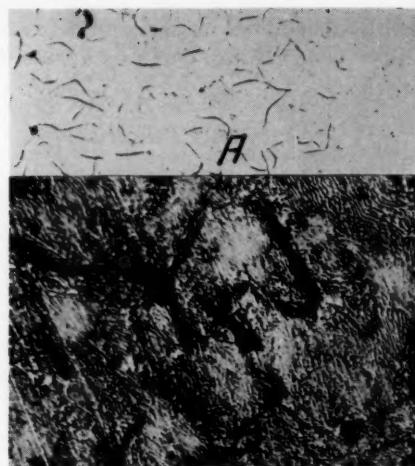


FIG. 7—A 2.75 PER CENT CARBON, 1.75 SILICON, 1.50 NICKEL, 0.60 CHROMIUM IRON, USED FOR DIES. THIS IRON FREQUENTLY HEAT TREATED FOR DEVELOPMENT OF SPECIAL PROPERTIES. A: SHOWING GRAPHITE DISTRIBUTION. UN-ETCHED, X100. B: SHOWING CHARACTERISTIC FINE PEARLITE. ETCHED IN 4% NITAL, X500.

find it desirable to pour castings for pressure work from a dense, tough iron, which receives a heat treatment to lower the Brinell hardness from 25 to 50 points, while changing the matrix only slightly, but which permits high-speed machining. In other cases we furnish castings which are machined as cast and then hardened and drawn, or which are hardened only.

45. In the production of heat-treated castings, the definite control of composition and temperature that we obtain with our

Table 4
PHYSICAL CHARACTERISTICS OF HEAT-TREATED IRONS

Heat No.	Yield Point, lbs. per sq. in.	Ultimate Tens. Strength, lbs. per sq. in.	Elongation in 2 in., per cent
806	46,405	67,525	13.00
	44,690	66,790	12.50
817	51,310	75,655	7.50
	51,630	76,075	7.50
873	71,400	100,370	4.00
	69,675	96,870	3.50

NOTE: Tests were made on A.S.T.M. standard 0.625-in. bar.

electric furnace enables us to establish routine methods for accurate production of the desired properties. This uniform control of the properties of these castings opens up a wide field of application akin to steel.

Promptness of Delivery

46. One outstanding advantage of our electric furnace foundry that should be mentioned before concluding, is our ability to make prompt delivery of castings. It is obvious that it is impossible to produce the wide range of compositions outlined by any method other than with the rocking electric furnace. In addition we have found it economical and expedient to produce in the electric furnace irons of ordinary cupola composition, which enables us to eliminate the delay incident in accumulating a sufficient tonnage of castings for a cupola heat. During such periods we often run 16 to 24 hours daily. In fact, there have been occasions where the furnace has been in operation continuously for more than 96 hours.

47. We are well aware that many applications of electric furnaces have been made for the production of specific castings such as piston rings, brake drums, cylinder blocks, camshafts and even crankshafts. However, in this paper we have attempted to show the practicability of using the rocking electric furnace for the production of a wide variety of castings to meet the requirements encountered in the operation of a general jobbing foundry, laying particular stress upon quality control and flexibility.

ORAL DISCUSSION

II. BORNSTEIN:¹ On page 523, paragraph 16, the author shows results, from 200 heats and obtains carbon contents ranging from 2.18 to 2.24 per cent. Carbon is within six points on 200 heats, including errors in chemical analysis. If the same iron was sent to a dozen different laboratories, there probably would be a carbon difference of at least six points reported.

In results on eight heats shown in Table 2 at the bottom of page 524, carbon content shows a variation throughout of only about six points, with the total carbon in the neighborhood of 3.0 per cent. A check of six points would be considered fairly good between two or three different laboratories. Mr. Culling probably is holding his metal right on the line, even better than the results show.

¹ Deere & Co., McLine, Ill.

H. S. AUSTIN:² The same thought occurred to me about the analyses. What I am wondering is, where can Mr. Culling get raw materials that make such results possible? I know the raw materials have been analyzed before use.

C. R. CULLING: We are quite particular about our raw materials which consist largely of steel scrap and cast borings. Clean steel scrap may be procured with practically no variation in chemical composition. In regard to borings, we do not use indiscriminately any type borings, nor do we mix different types in one storage bin. When a car load of borings arrives, we standardize that car first by making a careful chemical check of about ten samples from different parts of the car and then by melting several heats of the material. We check this analysis against that obtained from the originals. In this manner, we have standardized all the elements in that particular car of borings. Consequently, when the material is used, we know within chemical error what results we will get. We make a careful check in our remelt for chromium, nickel, and molybdenum as these elements tend to upset the results of the fracture bar test. All of our raw material is handled in the same manner.

CHAIRMAN J. W. BOLTON:³ There is another interesting angle not directly brought out in Mr. Culling's paper. Even 15 years ago, a laboratory around a foundry was unusual. Some foundries sent out samples once a week or once a month, and the check on mixtures was most casual. It is evident that we have progressed mightily in having to check each and every point as we go along. It would be utterly impossible to produce Mr. Culling's results without the most rigid technical control.

R. H. BANCROFT:⁴ Answering Mr. Bornstein's question, in producing 30 heats of considerably higher carbon content, approximately 3.65 per cent, in an indirect-arc furnace, we can show nearly a seven point carbon variation.

H. B. SWAN:⁵ Our type of melting is not comparable with Mr. Culling's. In our plant, we do not have to hold the analysis so close. We could not, because we use the Dulpexing process. Variations in the cupola are carried through the electric furnace. As far as strength and structure of one or two of these types of irons are concerned, we check him closely, particularly on the molybdenum iron with low carbon content.

CHAIRMAN BOLTON: It might be of interest to ask Mr. Culling how he checks his temperatures. The analysis can be very close and yet some variation in temperature at the time of operation would make a difference.

C. R. CULLING: Temperatures are checked carefully by an optical pyrometer which is calibrated daily. While we realize that temperatures measured by an optical pyrometer may not be absolutely correct, they are relevant and the results always will be uniform as long as the instrument is kept calibrated. Pouring and melting temperatures are both important and it naturally follows that uniform results require uniform practice.

² Campbell, Wyant & Cannon Co., Muskegon, Mich.

³ Lukenheimer Co., Cincinnati.

⁴ Perfect Circle Co., Newcastle, Ind.

⁵ Cadillac Motor Car Co., Detroit.

M. J. LEFLER:⁶ Do either of the gentlemen who just spoke take a preliminary analysis and do some doping of the heat to get the carbon content as uniform as is mentioned?

C. R. CULLING: In the standardization of raw materials, we do not attempt to definitely control our carbon in that manner. We check approximately for carbon because, in the use of borings, there is always some contamination by iron oxide and considerable variation in carbon content of the raw material. That is checked at the time the heat is made by a fracture test. If we find the carbon a few points low we add, by mathematical calculation, the proper amount of carbon, rock the furnace, pour and take another check by fracture bar. Should the carbon show high in the fracture test, we add the necessary amount of steel to reduce it.

H. I. THOMASMEYER:⁷ Did you check to find what your fuel costs were per ton compared to coke?

C. R. CULLING: The cost of current as a melting medium is considerably higher than the cost of coke. However, that is offset, to a certain extent, by the use of cheaper raw materials. The final result is comparable to what we call high grade cupola operation. The cost of producing certain grades of irons does not enter into the question, because we are unable to produce them in any other manner than with the electric furnace.

MEMBER: I would like to ask what method is used to deoxidize the bath. How much coke is used in calculating the charge? Do you add coke to the charge while melting or in the initial charge? How do you slag off, if there is any accumulation of slag? If there is no slag accumulation, what kind of raw materials do you use?

C. R. CULLING: Deoxidation is accomplished by the rocking action of the furnace coupled with the atmospheric reaction which takes place in the furnace simultaneously with superheating. This action probably is aided by the absence of any significant amount of slag on the bath.

In ordinary charges, the correct amount of coke is added with the cold charge so that it may be absorbed as the charge is melting. In other instances, such as described in the change of bath progressively, we add coke after a portion of the heat has been poured.

If the charge contains sprue or scrap castings with sand attached, there is naturally an accumulation of slag which is removed from the bath while it is in a viscous state before the superheating and deoxidizing stages. If the charge contains no material contaminated with sand, the amount of slag is negligible. We absorb in carbon content approximately 76 per cent of the coke by weight.

In regard to technical control, I have a letter where R. S. McPherran, Allis-Chalmers Mfg. Co., Milwaukee, commented on the fact that no definite reference had been made as to the necessity for chemical analytical control. I believe the discussion has brought out the fact that it is necessary to have careful analytical work to achieve definite, uniform results. We have a laboratory adjacent to our electric furnace and are extremely careful in our carbon determinations. All samples are run in duplicate and checked with U. S. Bureau of Standards samples. We also use in our

⁶ Oliver Farm Equipment Co., South Bend, Ind.

⁷ Syracuse Chilled Plow Co., Syracuse, N. Y.

routine checking a microscope such as was described in Roy M. Allen's paper* presented before this association in 1931.

A. C. MEYERS: Mr. Culling touched briefly on a great achievement, in my estimation. That is determining the approximate carbon content of metal by the fracture test. I have talked to him privately and it seems as if they have the procedure down to a fine point. They can tell to within a few points the carbon content of metal by the fracture test. I think that is a remarkable achievement.

* Williams & Co., Pittsburgh.

* R. M. Allen, *The Microscope as a Practical Aid in The Cast Iron Foundry*, TRANSACTIONS, American Foundrymen's Association, vol. 39, pp. 773-816, 1931.

Dynamic Properties of Steel Castings

By FRED GROTT,^{*} EAST CHICAGO, IND.

Abstract

This paper presents data to show the author's point of view that impact test results give more reliable information on the ability of cast steels to resist shock in service than do test results on tensile strengths and elongation. Grain structures as shown by micrographs indicate the probable impact values. To obtain best results as indicated by service tests and impact tests, heat treatments are of special importance. Similar steels, given different heat treatments, may give similar tensile strength results but vary widely in impact resistance. The author has shown the value of impact testing and he expresses the point of view that lack of ability to resist impact has been the cause of many so-called mysterious failures encountered in service.

1. The assembling of data for this contribution was prompted by references in papers recently written by R. A. Bull,^{1†} H. F. Moore,² Lorig and Williams,³ and Federico Giolitti.⁴
2. In Major Bull's paper, "Special Steels for Castings," the importance of impact testing was pointed out.
3. In Professor Moore's article on "Test Results and Service Value of Materials," references were made to brittle material not having plastic readjustment, prompting the question: "Is a material with very low impact classified as brittle?" The query arises because it has been found that certain materials with only 2 ft.-lbs. Izod (in other words, tending to break abruptly under dynamic stress) have from 25 to 35 per cent elongation and deform readily.
4. In the very comprehensive paper by Lorig and Williams on "Physical and Mechanical Properties of Some Well-Known Cast Steels," many data and comments are given on impact tests indicative of their value.

^{*} Director of Metallurgy, Continental Roll & Steel Foundry Co.

[†] Reference numbers as shown herein correspond with numbers in the Bibliography at the end of the paper.

NOTE: This paper was presented before one of the sessions on Steel Castings at the 1933 Convention of the American Foundrymen's Association.

5. In Giolitti's article on "Impact Tests in Specifications for Soft Steels," we note that the Italian State Railways technicians have been discussing new purchase requirements for metals and alloys. Satisfactory agreements were reached with no great difficulty on all points but one. The exception concerned the proposed inclusion of the impact test on notched bars into the standard specification, not only for special or heat-treated steel but for all ordinary untreated carbon steels, cast, rolled or forged.

Earlier Data Indicate Value of Impact Tests

6. These articles referring to impact tests are of considerable interest and indicate that such tests may be of great value in specifications and provide the explanation for many mysterious failures of metals. In order to assist in clarifying the subject, the author presents here the results of a series of researches conducted over a considerable period of time.

7. The first observations on impact testing of steel as recorded in this paper were made by R. G. Guthrie⁵ in 1918, when he was associated with the Curtis Aeroplane Testing Laboratory. The parts used for testing were airplane axles of 3.5 per cent nickel steel, given special heat treatments. Sometimes an axle failed rather abruptly when the airplane made a landing, and when these troubles occurred the parts were submitted to the laboratory for research. The specified heat treatment was to quench in oil from a temperature of 1550 degs. Fahr., then to draw at 1000 degs. Fahr. Fig. 1 shows the type of axle failure encountered.

8. In practically all cases, sections of the axles would give good physical tests for elongation and tensile strength. The impact values in Izod foot-pounds of steel samples from failed axles averaged 6 to 12 lbs. At that time there was no attempt to estab-



FIG. 1—AIRPLANE AXLE OF 3 1/2 PER CENT NICKEL STEEL, SHOWING ABRUPT FAILURE.

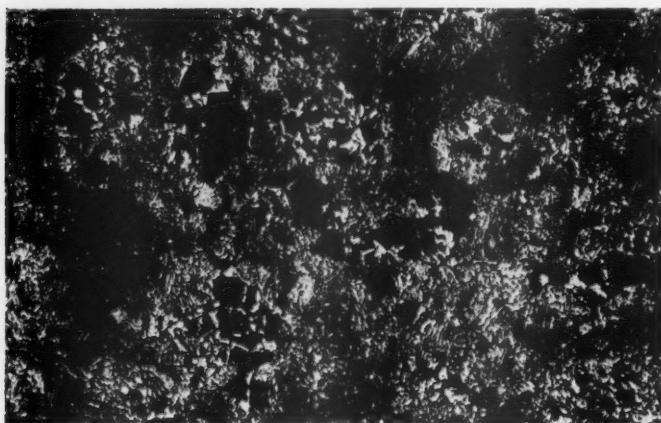


FIG. 2—MICROSCOPIC STRUCTURE OF AIRPLANE AXLE OF FIG. 1.

lish relationship between impact values and other physical properties. However, it was observed on steel of the same analysis, that impact values of 40 to 60 ft.-lbs. were often obtained, while the elongation and tensile strength values were approximately the same as those taken from failed parts.

9. The structures of the steel were variable, as shown in Fig. 2. This was evidence that the heat treatments were not constant. Therefore, the first effort was to get uniform results in heat treating, not solely with the intention of securing uniform impact results.

10. It is of interest to note that the failures were eliminated by establishing proper manufacturing practice and close inspection control. It appeared that the failure of these nickel steel axles was due to low impact value, in spite of the fact that tensile strengths and elongations checked with the original specifications. The corrective methods also probably resulted in high, uniform impact values, although it was not standard practice to check regularly for impact resistance.

Tank and Tractor Castings Must Resist Shock

11. Tanks and tractors are other types of equipment that encounter terrific shocks. In some cases the machine drops several feet into trenches and ditches, and in other cases it rams into walls and heavy obstacles. The construction and operation of tanks involved in the author's investigation were very similar to those of the large type of track-laying tractor familiar to many.



FIG. 3—CAST STEEL LINK FOR TRACTOR, SHOWING TYPE OF FAILURE OCCURRING PREVIOUS TO RESEARCH OF 1919-21.

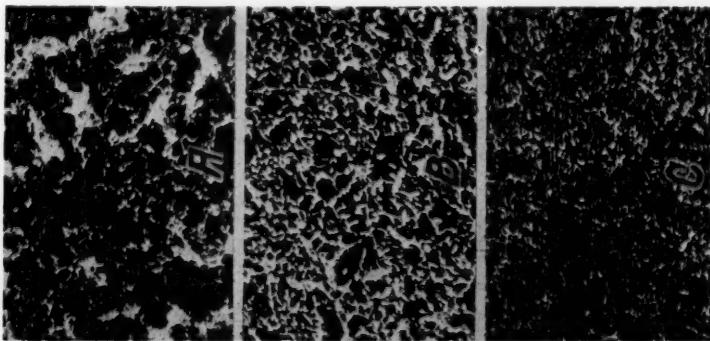


FIG. 4—*A*: SECTION TAKEN FROM CASTING; IZOD 4 FT.-LBS. X100. *B*: SECTION TAKEN FROM CASTING SHOWING IMPROVED PHYSICAL PROPERTIES OVER *A*; IZOD 12 FT.-LBS. X100. *C*: STRUCTURE OF CASTING GIVING GOOD SERVICE; IZOD 28 FT.-LBS. X100.

12. The fundamental part of this machine is the track. The machine functions somewhat like a locomotive traveling on a rail, with the difference that the rail is carried by the machine and laid down as required. Since the success or failure of this machine depends on the track, it was the track that received the most attention.

13. The track was made originally of cast steel links connected with pins, and Fig. 3 shows one such steel casting that caused the original trouble. This breakage occurred when the operator encountered a rather deep ditch without reducing his momentum, and numerous repetitions of this particular kind of failure occurred.

14. Col. Geo. D. Babcock,⁶ then in the Ordnance Department, U. S. A., requested that a research be conducted to correct the trouble. Preliminary observations showed the average steel casting to contain from 0.30 to 0.35 per cent carbon, 0.50 to 0.65 per cent manganese, 0.04 per cent sulphur, and 0.05 per cent phosphorus. The material was full annealed and had a structure as shown in Fig. 4-A. The physical properties averaged as follows:

Ultimate tensile strength, lbs. per sq. in.	78,000
Yield point, lbs. per sq. in.	47,000
Elongation in 2 inches, per cent	26
Reduction of area, per cent	38

15. The first attempt to correct the existing condition was to full-anneal and then normalize the casting. This resulted in a structure as shown in Fig. 4-B. The resulting physical properties averaged as follows:

Ultimate tensile strength, lbs. per sq. in.	82,000
Yield point, lbs. per sq. in.	54,000
Elongation in 2 inches, per cent	24
Reduction of area, per cent	42

16. These physical properties gave greatly improved service. However, research was continued along various lines with the result that two different procedures were perfected, one consisting of differential quenching and drawing, the other involving the use of special alloys. Carbon steel castings of composition similar to that previously mentioned were full-annealed, then quenched in warm water and drawn, and produced a structure as shown in Fig. 4-C. Physical properties averaged as follows:

Ultimate tensile strength, lbs. per sq. in.	81,000
Yield point, lbs. per sq. in.	58,000
Elongation in 2 inches, per cent	22
Reduction of area, per cent	50

17. These castings gave fine service. Experiments with medium manganese, chrome-nickel, and vanadium steel castings showed improved results, especially when given a treatment that resulted in a minimum of free ferrite. The author also devised a differential heat treatment for hardening the rails of these track links which gave better results than those obtained with other practices employed at that time.

18. Important steel castings other than the links used on the ordnance tractors mentioned, also subjected to tremendous dynamic stresses, were finally produced satisfactorily from steel having a structure similar to that shown in Fig. 6-C.

19. It is appropriate here to refer to Fig. 4, showing steels with very similar tensile strengths and elongations. The reduction

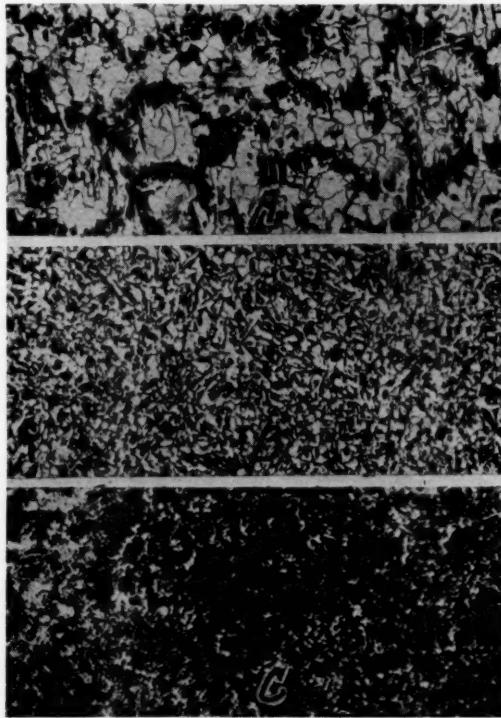


FIG. 5—*A*: MICROGRAPH OF STRUCTURE OF ANNEALED SIDE FRAME; IZOD 12 FT.-LBS. X100. *B*: MICROGRAPH OF STRUCTURE OF NORMALIZED SIDE FRAME; IZOD 22 FT.-LBS. X100. *C*: MICROGRAPH OF STRUCTURE OF SORBETIZED SIDE FRAME; IZOD 40 FT.-LBS. X100.

increased as the quantity of free ferrite was reduced. Resistance to dynamic stresses increased as the fineness of the structure increased. The structure as shown in Fig. 4-A was satisfactory for static application, but where dynamic service was encountered, abrupt failure resulted. It also was observed that an added element, such as nickel or vanadium, seemed to increase the toughness or resistance to shock.

High Impact Values Needed for Freight Car Equipment

20. The following comparative data obtained by the author on steel in side frames for freight-car trucks may be of interest. Chemical composition of the castings averaged as follows:

	Per Cent.
Carbon	0.29
Manganese	0.70
Silicon	0.38
Sulphur	0.03
Phosphorus	0.03

21. Full-annealing resulted in a structure like that shown in Fig. 5-A, and in the following physical properties:

Ultimate tensile strength, lbs. per sq. in.	65,000
Yield point, lbs. per sq. in.	38,000
Elongation in 2 inches, per cent	29
Reduction of area, per cent	40
Izod, ft.-lbs.	12

22. Fig. 5-B shows the structure of material having the following physical properties after full-annealing and normalizing:

Ultimate tensile strength, lbs. per sq. in.	67,000
Yield point, lbs. per sq. in.	40,000
Elongation in 2 inches, per cent	26
Reduction of area, per cent	45
Izod, ft. lbs.	22

23. Fig. 5-C shows the structure of material having the following physical properties after full-annealing, quenching and drawing:

Ultimate tensile strength, lbs. per sq. in.	70,000
Yield point, lbs. per sq. in.	46,000
Elongation in 2 inches, per cent	25
Reduction of area, per cent	52
Izod, ft.-lbs.	40

24. This contrast of physical properties in railroad castings is of interest in that the tensile strengths and elongations in the three samples are practically the same, yet the impact values vary from 12 to 40. Data subsequently obtained showed Izod values ranging from 4 to 12 in steel the sections of which were closely

calculated, having a tendency to fail when subjected to a violent blow. Fig. 5 shows structures from material of the same analysis.

25. Castings with structures similar to that of Fig. 5-C, when used for tests approximating road conditions, appear to last indefinitely so far as fatigue or reversal of stress is concerned. Observations in general thus far seem to show that when the ferrite is in solution and etching reagents indicate a sorbitic structure, performance under dynamic stresses is highly satisfactory.

Mysterious Failures Traced to Poor Dynamic Properties

26. The accumulated evidence seemed to indicate, at this point in the investigation, that many unknown failures are traceable to poor dynamic properties, and that good resistance to dynamic stresses results from a certain fineness of structure, or ratio of free ferrite. The thought was advanced by J. Campbell⁷ that an extra element, such as vanadium or zirconium, might act to prevent certain slip movements in the crystalline structure of the steel, and thus provide a toughness not characteristic of straight carbon steels.

27. The general data obtainable from fatigue and tensile tests did not seem to give a satisfactory solution to the problem. From extensive service observations it appeared that sudden blows were always the source of the trouble, rather than straight static loads or reversal of stress loading. It was very noticeable in some cases that metal that would take extensive distortion under static conditions would fail quickly under a sharp blow.

28. Accordingly, data were tabulated after research along the following lines:

- (a) To determine the effect of grain size.
- (b) To determine the effect of free ferrite.
- (c) To determine the effect of extra elements.
- (d) To determine if the impact test is a gage of dynamic service.
- (e) To determine if impact is indicated by structure.
- (f) To observe the relation, if any, of impact to other physical properties.

Investigation of Reclamation and Dredging Castings

29. The service encountered by reclamation and dredging machinery requires material of the very best design, composition and treatment. Among such parts are cast steel sprockets similar to those used in track-laying machines. It is the practice to make the hub and spokes tough and machinable with a Brinell

hardness about 190, while the teeth are given a hardness of from 400 to 500 Brinell to resist abrasion.

30. In this connection, attention is directed to Fig. 6. The analyses of the steel used for the four samples shown were practically the same, being approximately 0.35 per cent carbon, 1.55 per cent manganese, 0.35 per cent silicon, 0.04 per cent sulphur, and 0.03 per cent phosphorus. The fineness of structure is a gage of the impact value, as the steel illustrated in Fig. 6-A showed 4 ft.-lbs.; that in Fig. 6-B, 12 ft.-lbs.; that in Fig. 6-C, 33 ft.-lbs.,

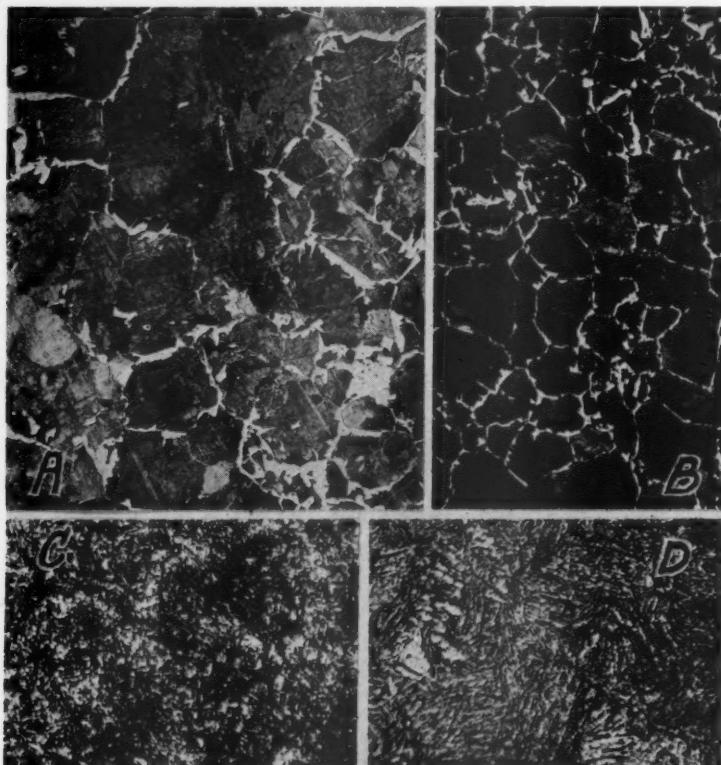


FIG. 6—STRUCTURES OF SIMILAR STEELS (CARBON 0.35, MANGANESE 1.55), SHOWING EFFECT ON GRAIN SIZE AS RELATED TO IMPACT RESULTS BY DIFFERENT HEAT TREATMENTS. A: IZOD 4 FT.-LBS., NORMALIZED FROM 1700 DEGS. FAHR. X100. B: IZOD 12 FT.-LBS., ANNEALED AT 1650 DEGS., NORMALIZED FROM 1600 DEGS. X100. C: IZOD 33 FT.-LBS., ANNEALED AT 1650 DEGS., RAPID NORMALIZE FROM 1550 DEGS. AND DRAWN AT 1200 DEGS. X100. D: IZOD 10 FT.-LBS., SAME ANNEAL AS C BUT QUENCHED FROM 1550 DEGS. INSTEAD OF RAPID NORMALIZING. X100.

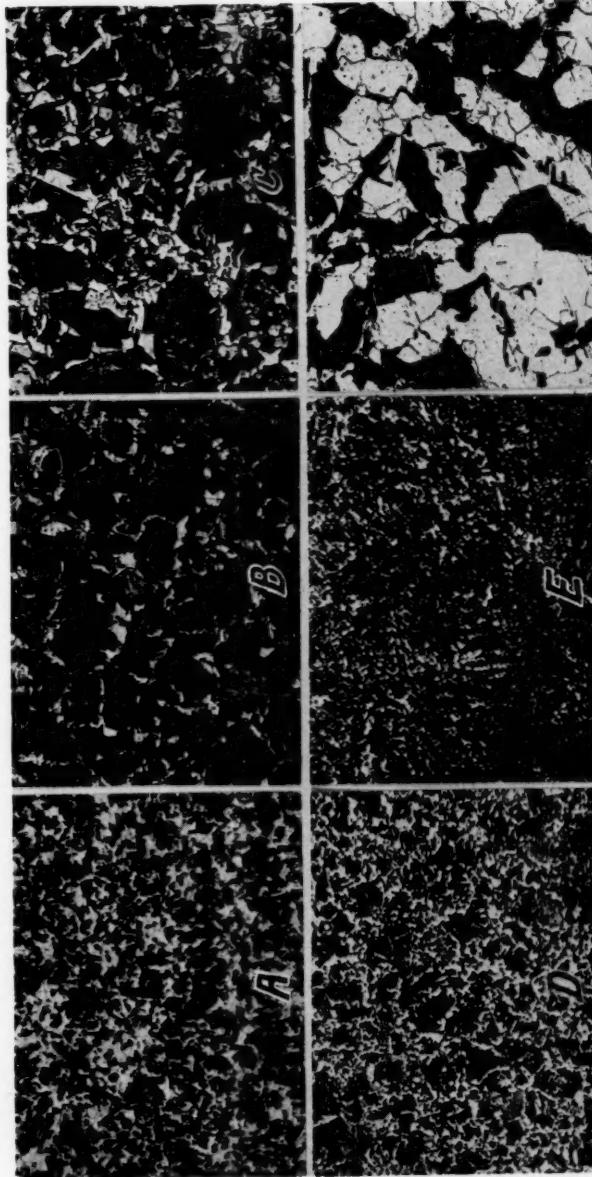


FIG. 7—EFFECT OF VARIOUS QUANTITIES OF FERRITE ON STEELS WITH SIMILAR TENSILE STRENGTH AND ELONGATION. A : IZOD 25 FT.-LBS. ; 0.25 CARBON, 1.40 MANGANESE ; 1600 DEGS. FAIR. AIR QUENCH, DRAW AT 1250 DEGS., 82,000 LBS. TENSILE, 30% ELONGATION. X100. B : IZOD 6 FT.-LBS. ; 0.34 CARBON, 1.40 MANGANESE ; 1675 DEGS. NORMALIZED OR COOLED IN STILL AIR ; 86,000 LBS. TENSILE, 20% ELONGATION. X100. C : IZOD 5 FT.-LBS. ; 0.34 CARBON, 1.54 MANGANESE ; 1700 DEGS. NORMALIZED IN FURNACE WITH DOORS UP ; 84,000 LBS. TENSILE, 28% ELONGATION. X100. D : IZOD 30 FT.-LBS. ; 0.30 CARBON, 1.40 MANGANESE, 0.10 VANADIUM ; 1650 DEGS. ANNEALED, 1550 DEGS. NORMALIZED IN STILL AIR, AND DRAWN AT 800 DEGS. ; 92,000 LBS. TENSILE, 26% ELONGATION. X100. E : IZOD 60 FT.-LBS. ; 0.32 CARBON, 1.20 MANGANESE, 0.60 NICKEL ; 1650 DEGS. ANNEALED, 1600 DEGS. DRAW ; 91,000 LBS. TENSILE, 30% ELONGATION. X100. F : IZOD 8 FT.-LBS. ; 0.32 CARBON, 1.40 MANGANESE ; 1700 DEGS. ANNEALED, 1600 DEGS. ANNEALED AND SLOW COOL IN FURNACE ; 84,000 LBS. TENSILE, 28% ELONGATION. X100.

and that in Fig. 6-D, only 10 ft.-lbs. In the material shown in Fig. 6, A, B and C, tensile strength and high elongations were practically equivalent, while the steel shown in Fig. 6-D had poor elongation and high strength.

31. Fig. 7 shows the effects of various quantities of free ferrite. In Fig. 7, E and F show extremes, and although the tensile strengths and elongations are similar, the impacts are 60 and 8, respectively.

32. Fig. 7-D shows the effect of adding vanadium. It is apparent that under ordinary conditions a certain fineness results from the use of this element, with corresponding increase of impact resistance. It also has been shown by other researches, es-

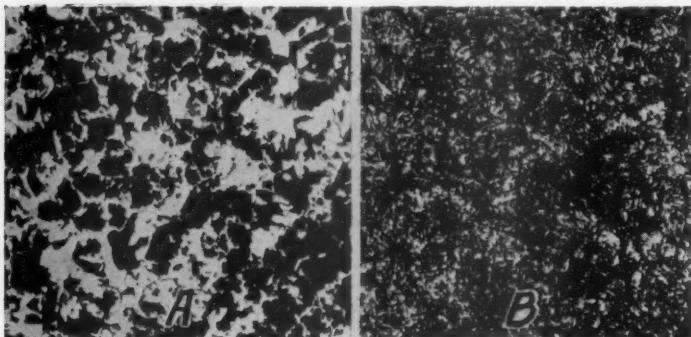


FIG. 8—A : STRUCTURE OF ANNEALED SPROCKET ; IZOD IMPACT 8 FT.-LBS. ; 1650 DEGS. FAHR. ANNEALED ; 85,000 LBS. TENSILE, 28% ELONGATION. X100. B : STRUCTURE OF DYNAMIC SPROCKET ; IZOD IMPACT 52 FT.-LBS. ; 1650 DEGS. ANNEALED, 1550 DEGS. RAPID NORMALIZE, 1250 DEGS. DRAW ; 88,000 LBS. TENSILE, 29% ELONGATION. X100.

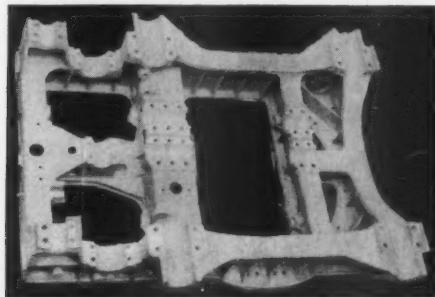


FIG. 9—LARGE CRAWLER FRAME HAVING UNIFORMLY HIGH DYNAMIC PHYSICAL PROPERTIES ; IZOD 55 FT.-LBS. (AVERAGE OF 10 HEATS)

pecially those reported by W. C. Hamilton⁸ and Jerome Strauss,⁹ that the addition of vanadium to medium pearlitic steel approximately doubles the impact values and gives a finer structure than is obtained without the element.

33. It is further demonstrated, as we contrast Figs. 8-*A* and 8-*B*, that in some cases the micrograph will indicate the impact value. In these cases of the same analysis, the steel shown in Fig.

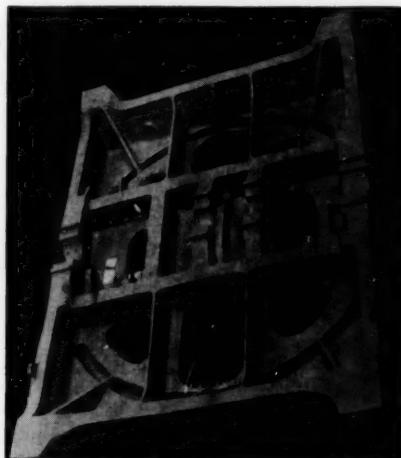


FIG. 10—EXCAVATING MACHINE BASE WITH SPECIAL DYNAMIC PROPERTIES; IZOD 40 FT.-LBS.

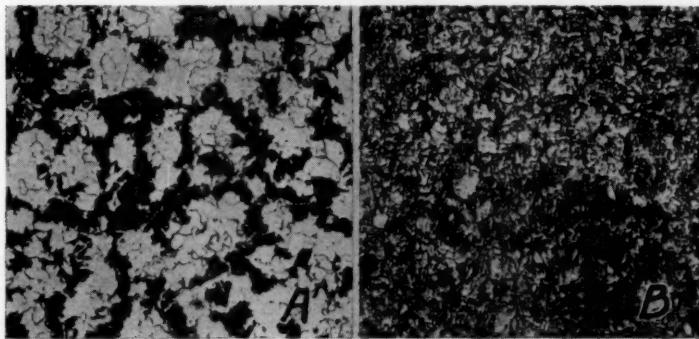


FIG. 11—*A*: STRUCTURE OF CASTING; IZOD 4 FT.-LBS.; 1675 DEGS. FAHR. FURNACE ANNEALED; 86,000 LBS. TENSILE, 27% ELONGATION. X100. *B*: STRUCTURE OF CASTING AFTER PROPER HEAT TREATMENT; IZOD 58 FT.-LBS.; 1675 DEGS. ANNEALED, 1550 DEGS. RAPID NORMALIZE, 1250 DEGS. DRAW; 88,000 LBS. TENSILE, 29% ELONGATION. X100.

8-A had 8 ft.-lbs., while the steel shown in Fig. 8-B had 52 ft.-lbs. Izod values.

34. Figs. 9 and 10 illustrate large castings made of special steel treated to give controlled impact value and other special physical properties. They show what can be done through special efforts to produce large designs with minimum weight and high dynamic value. Practically no welding whatsoever is permitted on these castings, so that uniformity of results is guaranteed. It has been demonstrated that impact values of welded structural parts are variable as contrasted with a well-designed casting properly made



FIG. 12—A: SAME STRUCTURE AS SHOWN IN FIG. 11-A, BUT X1000; IZOD 4 FT.-LBS. B: SAME ANALYSIS AS A ABOVE, BUT HIGH IMPACT VALUE; IZOD 55 TO 60 FT.-LBS. X1000.

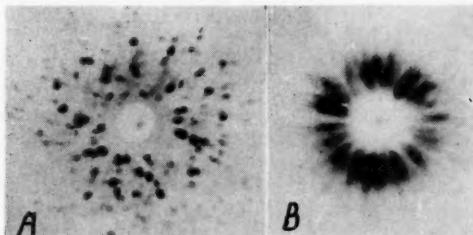


FIG. 13—A: RADIOPHOTOGRAPH OF SPECIMEN SHOWN IN FIG. 12-A. B: RADIOPHOTOGRAPH OF SPECIMEN SHOWN IN FIG. 12-B.

and treated. Reliability in these large designs is of maximum importance.

35. Changes in manufacturing procedure were effected for these parts. The standard practice had consisted of making the castings from a manganese pearlitic steel, well annealed, to give the required high qualities. This method resulted in very good bending properties, accompanying an elongation of about 30 per cent and a tensile strength averaging 78,500 lbs. per sq. in. The impact value, however, was 4 ft.-lbs. in steel having a structure as shown in Figs. 11-A and 12-A, the latter showing the condition at 1000 diameters magnification.

Slow Loading Versus Shock

36. It is clearly indicated that the constituents are completely divorced and that their ability to cohere must be at a minimum. However, it is apparent that if loading is slow, great elongation or plastic readjustment will occur, while the material might fail abruptly when subject to shock.

37. Fig. 13-A reproduces a radiograph showing a definite pattern for this structure. Since uniform dynamic properties were essential for these large castings, a practice was developed that resulted in a structure as shown in Fig. 11-B, in material having an Izod impact value of 55 to 60 ft.-lbs. The structure at 1000 diameters as shown in Fig. 12-B is not massive, as it is in Fig. 12-A. Fineness and uniform distribution are noticeable. The radiograph illustrated in Fig. 13-B shows a distinct pattern, as contrasted with that in Fig. 13-A (from W. G. Praed¹⁰).

38. The last section of this paper, mentioning large castings, is contributed to show that dynamic properties are controllable when correct practices are observed, and that the impact value of the steel is the best criterion on which to base the service value of such designs.

Summary

39. In summarizing researches engaging the author's attention to this subject, the following comments may be of interest:

(a) No attempt was made to accumulate data with respect to impact results on abnormal or improperly made steel.

(b) Inclusions of various types have much to do with impact values. Angular inclusions are more objectionable than round inclusions. The location of the inclusion with respect to the grain boundary has much to do with the ability of the material to hang together (F. Sutherland¹¹).

(c) Probably one of the most important features observed microscopically is the condition of the ferrite and grain refinement, as shown in Figs. 12-A and 12-B.

(d) No noticeable relationship was observed between tensile strength and impact value. (Tensile strength has a relationship to fatigue resistance, as observed by other investigators.)

(e) There seems to be a relationship between the yield point and impact values. In observed cases an increase of yield point was accompanied by increase of impact resistance.

(f) The elongation had no noticeable relation, in general.

(g) In practically all cases the impact value varied with the reduction of area value, good reduction attending good impact resistance.

(h) There was no connection between Brinell hardness readings and impact values.

(j) Our observations show that fatigue resistance varies with impact resistance, in that failure from repetitions of a measured stress are greatly reduced when materials have high impact value. This is an interesting feature, as fatigue value has a relationship to tensile strength whereas impact value has none. Still, by our observations there is a distinct relationship between impact resistance and fatigue resistance.

40. In conclusion, the author has attempted to show the value of impact testing of steel castings. Experiments on material for several purposes (aircraft, tractor, railroad and excavating machinery) have been mentioned, as in each case the application of impact testing resulted in practices which gave good dynamic properties that eliminated many troubles. There is no doubt that lack of ability to resist impact has been the cause of many so-called mysterious failures encountered in service.

41. Reference has been made to writers of valued papers relating to special physical properties of steel. In those contributions were striking paragraphs which indicate that impact testing of materials requires further attention and elaboration. The

author of this paper has briefly presented information obtained in some of his researches with an idea of increasing the general knowledge on the subject, and to indicate results obtained through imparting dynamic properties to cast steel by information obtainable from the impact test.

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WRITTEN DISCUSSION

D. C. ZUEGE:¹ It is recognized generally that steels of the intermediate-manganese type, either with or without the addition of other alloys, and most other alloy steels, must be given a draw treatment after normalizing if high impact values are required. In many cases, castings are given triple treatments in which two of the heatings have been to temperatures above the critical range, with highly beneficial results. It has been our experience, based upon test coupons attached to small and medium sized castings, that a double normalize and draw has given better results in securing the desirable, fine grained, sorbitic structures and high impact values, than the full anneal, normalize and draw treatment.

With reference to the relation of impact values to other physical properties, the statements made by the author in paragraphs d, e and f, page 552, should be qualified somewhat since they hold only where the microstructures or drawing temperatures are similar. A specimen which is martensitic or troostitic in structure generally has a *lower* impact

¹ Silver Steel Casting Co., Milwaukee.

value than a sorbitic structure but has a much higher yield point. The impact values of alloy cast steels which have been normalized are increased considerably by draw treatments and, in many instances, increase as the draw temperature is increased, notwithstanding the fact that such treatment decreases the yield point and tensile strength and increases elongation and reduction of area. Furthermore, since these changes in physical properties occur at draw temperatures below the critical range, they occur without any changes in the free ferrite distribution. Therefore, it appears that the relation between impact values and other physical properties is more involved than is indicated in the author's summary.

The author rightly emphasizes the necessity for a method of evaluating the shock resisting properties of engineering materials. Present methods of impact testing have enabled us to obtain relative information on this subject. However, because of the many variables introduced and the lack of standardization in impact testing, caution must be used in the interpretation of impact values. As a typical example, our laboratory files show a 24.9 ft. lb. Charpy value on an intermediate-manganese-nickel steel, whereas the Izod value on the same steel was 42 ft. lbs. Another steel, having 5 ft. lbs. Charpy when one type of notch was used, showed 11.5 ft. lbs. when a second commonly used type of notch was employed.

These examples have been included not for the purpose of condemning a test as valuable as the impact test, but rather to show that until a greater degree of standardization has occurred, impact test values are only relative and can be interpreted intelligently only when all conditions surrounding the tests are known. A steel showing, for instance, 40 ft. lbs. Izod is not necessarily a tougher steel than one showing 20 ft. lbs. Charpy.

ORAL DISCUSSION

F. GROTT²: It was not my purpose to discuss heat treating because that is a subject for a paper in itself. However, let us refer to the castings that have been shown. We have found that the practical production process is to anneal, rapidly normalize, and draw, because it is necessary to keep the constituents in solution. In ordinary foundries which have the car type annealer, it is essential that the castings be separated properly to heat correctly, also that cooling be uniform and rapid. Regarding ability to obtain special dynamic properties, there is no question of obtaining results from these designs if the particular features mentioned are watched.

I don't know whether I have answered the question completely, but castings must be well distributed so that cooling will be accentuated. Then the castings must be drawn. We do not consider double normalizing or annealing and normalizing sufficient. In general, the material is heated to a fairly high annealing temperature to refine the casting structure and reduce danger of cracking, followed with a lower temperature above the critical range, thus handled so as to retain important constituents and finally drawing back.

R. A. BULL³: When the term normalize is used, not all the facts

² Continental Roll & Steel Foundry Co., East Chicago, Ind.
³ Consultant on Steel Castings, Chicago.

always are known. Steel supposedly normalized may get many kinds of treatment. The ordinary practice of normalizing is to pull a truck out of a hot oven. It may be loaded five or six feet high. The truck may be quite wide. It generally is assumed that the material all is normalized properly. If the treatment of the casting in the bottom of the truck is considered, it must be realized that the material is not normalized.

When conclusions are drawn about various methods of heat treatment, such items must be taken into consideration. Sometimes in commercial practice, it is difficult to standardize all conditions in respect to every individual casting that may be under consideration.

Mr. Zuege's comments relating to the advisability of interpreting impact test values need a lot of thought. It is extremely unfortunate that we do not have one impact test which is accepted generally as standard. If either the Charpy or the Izod test is used in a plant and its customers use and think in terms of either of these tests, comparison may not have gone far enough. The manner in which the slot is made and the accuracy with which the tooling is done, exert a tremendous influence on the results of a particular test. Slight tool marks, that may not be obliterated entirely by the finest polishing operation, unquestionably have a decided influence on the resistance to the impact. That holds true to a greater extent in the more complex alloy steels as compared to plain carbon steel.

H. H. BLOMJO⁴: I have done quite a bit of work on tensile impact testing of cast steel and I am not as sure of its value as I was formerly. The tensile impact test does not show any preference for heat treatment that is not shown by the static tensile test. I redesigned the standard tensile impact bar so that it had the same relation between the diameter and gage length that the standard static tensile bar has, and find that the relation between elongation and reduction of area is in the same ratio for the tensile impact bar as for the static tensile bar, and almost exactly the same percentage. If your elongation is 20 per cent in the static test, it will be close to 20 per cent in the impact test.

F. GROTT⁵: It is not my idea to recommend the incorporation of impact testing in all specifications. It is a valuable item on certain castings, without a doubt, but the purpose of the paper is *not* to advocate impact in all specifications. If that is done, immediately the customer might ask for impact tests on all castings with the corresponding cost increase. As this discussion goes on, I wish to say that the endeavor is to bring out the importance of dynamic steel and not necessarily to incorporate the impact test in our specifications in general.

In reply to Major Bull, who mentioned machining and other imperfections, that applies also to the tensile bar. These bars must be reasonably correct. I do not advocate Izod over Charpy or vice versa. Either is an indication of the dynamic properties. We have found that the approximate ratio of Izod to Charpy is about 1.41. We have contrasted quite a large number of results. Multiply the Charpy by that ratio and the result is Izod.

⁴ Minneapolis Electric Steel Castings Co., Minneapolis.

A. W. LORENZ:⁵ Where Charpy and Izod tests are concerned, a certain relationship is found after running a number of tests but the type of notch must be predetermined and standardized. I cannot say much about the statements made on tensile impact results. Some of the statements sounded strange. The Watertown Arsenal, which has the only large Charpy impact machine in the country, did not place any confidence in the tensile impact test on the small machine because the size of the bar is roughly only one-eighth of an inch in diameter. The Arsenal has a machine of ten times the capacity which holds a bar of standard size. I know they are very much pleased with the result they get from that type of testing, but I wouldn't be disposed to trust results on cast steel on one-eighth inch diameter bar.

J. STRAUSS:⁶ A number of years ago I had grave misgivings regarding the use of a V-type notch in an Izod test bar. Some types, through experience, have been discarded. For a long time, I so mistrusted the V-type bar that I used the drilled notch, and began to think a little more about the subject. Petrenko, of the bureau of standards studied the notch question and reached some interesting conclusions. These are included in an article published about eight years ago.

The first point is that the degree of sensitivity to variation in the form and surface of the notch is dependent upon the character of the material. This has been confirmed in my own work. Brittle materials are more sensitive to variation in the notch than are the tougher materials. For that reason, some of the conclusions which Mr. Grotts has presented are independent of consideration of small differences in notch form and surface.

If a V-notch of the type standardized by the British Engineering Standards association is machined in an Izod test bar with a form cutter in a single cut, it will not make a good bar and it is difficult to check results within the same laboratory and with those of other laboratories. If the notch is made with several cuts, the last being three or four thousandths of an inch thick, small variations in surface due to the small scratchings resulting from the machine tool will not be appreciable. We have checked that point carefully. We have checked our own Izod bars with Izod bars prepared elsewhere and also with tests by other laboratories and the degree of agreement is satisfactory.

If an Izod test is made on a steel casting and variations among different laboratories and different methods of preparation of the test bar are from 55 to 60 ft.-lbs., how much difference does it make? However, if the material has an Izod value of 15 ft.-lbs., the same variation in surface condition in the notch of the test piece may cause a variation of from 10 to 20 ft.-lbs. That is an appreciable difference.

We prepared test pieces in which the notch was ground by emery wheels of several different degrees of hardness and also machined by the use of a form cutter. We find that any carefully thought out method is satisfactory, provided the radius at the base of the notch is reasonably uniform.

⁵ Bucyrus Erie Co., Milwaukee.

⁶ Vanadium Corporation of America, Bridgeville, Pa.

WRITTEN REPLY BY AUTHOR

F. GROTT: I wish to thank the various individuals who have presented discussions for their constructive and interesting ideas on impact testing with relation to gaging the dynamic properties of steel castings. There is much to be gained by further investigation and discussion. The point that castings can have extensive deformation properties under static conditions, and yet fail suddenly when subject to abrupt working conditions especially should be elaborated.

A private communication from Mr. Batty asks if I consider an impact test, with a Brinell reading, sufficient criteria for acceptance of a casting. I do for some applications. In general, I would recommend it as an addition to regular tests as it is the measure of a particular property not shown conclusively by the tensile, elastic limit, elongation or reduction tests. There seems to be a relationship to the reduction of area and yield point. A Brinell result, with an impact test, may be better for some dynamic designs than figures showing strength and ductility.

The question of surface variation, as brought out by Major Bull and Mr. Zuege, has been answered by Mr. Straus. With regard to standardization of machining methods for impact test bars, there is no doubt about the importance of such a procedure.

Symposium, Tests and Specifications for Cast Iron

Relationship of A.S.T.M. and A.F.A. in the Development of Tests and Specifications

BY H. BORNSTEIN,* MOLINE, ILL.

In the field of cast iron for many years there has been excellent cooperation between the A.F.A. and American Society for Testing Materials. The A.F.A. has encouraged research in respect to manufacture and the A.S.T.M. has devoted itself to testing and specifications. These fields are related, and this relationship has been recognized in the committee work of the two organizations.

In 1922 the officials of A.F.A. and A.S.T.M. endeavored to secure closer cooperation between the two associations on matters of mutual interest. A joint committee was appointed and the following recommendations were made:

- (a) That the A.F.A. be elected to membership on those A.S.T.M. committees dealing with specifications and methods in which the A.F.A. is interested.
- (b) That the chairmen of the A.F.A. committees be designated as the A.F.A. representatives on the subcommittees of the appropriate A.S.T.M. committees.
- (c) That the A.F.A. committees function by reviewing A.S.T.M. specifications as they stand, either recommending endorsement by the A.F.A. or submitting criticisms for transmittal to the A.S.T.M. committees. On the A.S.T.M. tentative specifications, endorsement should be deferred until their adoption as standard, and that in the meantime any criticism of such specifications should be transmitted to the A.S.T.M. committee.
- (d) That with reference to publication by the A.F.A. of A.S.T.M. specifications, that publication take the form of an endorsement of such specifications, which will be printed as A.S.T.M. specifications.

* Director, Testing and Research Laboratory, Deere & Co.; Chairman A.F.A. Gray Iron Division, and Chairman Committee A-3 on Cast Iron of the American Society for Testing Materials.

NOTE: This and the seven following papers constitute the "Symposium on Tests and Specifications for Cast Iron" presented at a Joint Meeting of the A.F.A. and the American Society for Testing Materials held as a part of the 1933 Convention of the American Foundrymen's Association.

The above recommendations were officially approved by both A.F.A. and A.S.T.M. in 1922. During the past year (1932) an arrangement was made between the two societies whereby A.S.T.M. specifications, when approved by the A.F.A., will carry a notice that such A.F.A. approval has been given.

In respect to A.S.T.M. Committee A-3 on Cast Iron, this plan has worked very well. A number of members of the A.F.A. Committee on Gray Iron Castings are also members of Committee A-3. This has resulted in the closest cooperation on problems of mutual interest, and it has prevented duplication of effort. It is hoped that the pleasant and profitable relations between the two organizations will long be continued.

Significance of Testing Cast Iron and Limitations of Testing

BY H. BORNSTEIN,* MOLINE, ILL.

The user of a casting is interested primarily in the performance of this casting in service. Our testing procedure should have as an aim the securing of information which may be used in the estimation of the suitability of the casting for service. Too frequently we lose sight of the function of testing, and as a result many tests are made which have little or no relationship to the value of the casting for its particular service. Since iron castings are used for many types of service, different types of tests are required.

These tests include chemical analysis, various physical tests such as tensile, transverse, compression, shear, hardness, impact, etc., and examination of fracture and structure. Occasionally special tests for machinability and wear are employed.

Chemical Analysis for Production Control

Chemical analysis is valuable to the producer of castings as a means for control of his product. It is of lesser value to the user because two castings of similar chemical analysis may vary widely in physical properties. For this reason specifications are now stressing physical properties rather than composition. Occasionally, alloying elements are added to secure certain desired properties, and it may be difficult to describe these properties in terms of physical test results. In such cases the specifying of these elements may be desirable.

The consumer is concerned with the suitability of his casting for service, and he should not hamper the producer with specifications for composition which may adversely affect this service. There are many such specifications today, and the foundryman can do his part by pointing out unnecessary or inconsistent specifications to the consumer.

* Director, Testing and Research Laboratory, Deere & Co.

NOTE: This paper was the second of eight contributions presented at a Joint Meeting of the A.F.A. and the American Society for Testing Materials held as a part of the 1933 Convention of the American Foundrymen's Association

Service Tests

The best test of a casting is a service test. Obviously, we cannot use this as an acceptance test. However, where strength is a factor, the testing of an occasional casting for strength is recommended, provided the casting lends itself to such a test. Also, the proper size of test bar should be selected; then a correlation might be established between the casting and the test bar.

No standard tests have been worked out for wear resistance. There are many types of wear, and an iron which may give good performance under one set of conditions may be poor under other conditions. Frequently, tests for hardness and structure may indicate the results to be obtained.

Knowledge of Test Limitations Important

It should be pointed out that tests on cast iron are valuable if we realize the limitations of testing and provided sufficient care is taken in the interpretation of test results.

Various Tests for Cast Iron

BY R. S. MACPHERRAN,* MILWAUKEE

1. Transverse Test

The transverse test is the most widely used test for cast iron in this country. It is a popular test because the test bar is not machined and the test is easily made. The breaking load and deflection are usually determined, giving information which is of value and which may be translated into other terms. In some cases the deflection at a certain load is determined in order to get a measure of the stiffness of the metal.

The modulus of rupture may be calculated from the results of the transverse test. The modulus of rupture is equal to

$$\frac{P l e}{4 I}$$

Where P = breaking load in pounds.

l = distance between supports in inches.

e = distance from neutral axis to outer fiber.

I = moment of inertia.

For the three sizes of transverse test bars in the A.S.T.M. Tentative Specifications for Gray Iron Castings, A48-32T, the modulus of rupture may be obtained by multiplying the breaking load by the factor (equal to value of $l e / 4 I$) in the following table:

Test Bar		
Diameter.	Distance between supports.	Factor.
0.875 inches.	12 inches.	45.61
1.2	18 inches.	26.53
2.0	24 inches.	7.64

The reader is referred to the Symposium on Cast Iron to be presented at the A.S.T.M. meeting June 26, 1933, for a more extended discussion.

2. Tensile Test

The tensile test for cast iron has gained rapidly in importance. Tensile strength figures usually mean much more to the engineer than do the results of the transverse test.

* Chief Chemist, Allis-Chalmers Mfg. Co.

NOTE: This paper was the third of eight contributions presented at a Joint Meeting of the A.F.A. and the American Society for Testing Materials held as a part of the 1933 Convention of the American Foundrymen's Association.

There is no definite relationship between tensile test results and transverse test results. For a given iron, a fairly good relationship may be established so that in a particular foundry the results of transverse test can be translated into approximate tensile figures. However, this relationship will vary depending on the type of iron used.

The tensile test bar may be machined from the broken halves of the transverse test bar or may be machined from a separately cast tensile test bar.

The cost of machining is one of the objections advanced to the use of the tensile test bar. A number of attempts have been made to reduce the amount of machining required. Care should be taken to obtain axial loading in the tensile test, otherwise the results will be of little value.

3. Hardness Tests

Hardness tests are frequently made on cast iron. The Brinell test is most popular. The Rockwell hardness test is also employed, using the *B* scale with steel ball for gray irons and the *C* scale with diamond penetration for white irons. The sclerometer may also be used for white irons.

For particular irons, relationships may be established between hardness and strength, machinability, etc. However, no general relationship may be used.

4. Impact Tests

Shock or impact tests are made on cast iron. Various types of testing machines are employed and no standard method has been adopted. For a detailed report on impact testing of cast iron, reference should be made to the 1933 report of Committee A-3 of the American Society for Testing Materials.

5. Chill Test

The chill test is used in many foundries as a control test, and is quite valuable. No standard test has been adopted, but each foundry uses a size of bar suitable to its particular work. A definite relationship may be established in the foundry between the chill test results and the metal in the casting.

6. Fatigue Test

The fatigue test is not generally used for cast iron. Prof. J. B. Kommers made a large number of fatigue tests on cast iron

and some of this work is reported in his paper on "Fatigue and Static Properties of Some Cast Irons," *A.S.T.M. Proceedings*, 1928, page 174. The endurance limit in fatigue is about 50 per cent of the tensile strength.

7. Compression Test

Cast iron has a high value in compression. The compression test is made by loading a one-inch cube until failure occurs. The compression strength of cast iron ranges from about $2\frac{1}{2}$ to 5 times the tensile strength. In general, the ratio is higher for the weaker irons and lower for the stronger irons.

Correlation of Test Bar and Casting

BY R. S. MACPHERRAN,* MILWAUKEE

After making tests on test bars, the next thing is to translate these values into terms of the iron in the casting itself.

The correlation of test bar and casting has long been the dream of every foundryman and metallurgist. It always has been recognized that the strength of the test bar itself, in pounds per square inch, did not represent the strength of the same iron when poured into a casting of any size. This is due to the differences in the speed of cooling, internal stresses, soundness, etc. The slower the cooling, naturally the softer the iron.

It is an old saying that any plain iron is soft if cooled slowly enough and any plain iron is hard if cooled fast enough. In fact, the cooling conditions of any ordinary casting have more effect on its physical properties than does the chemical analysis.

Many efforts have been made by various men and organizations to estimate the strength of the iron in the casting from the physical tests made on the test bar. It seemed to the writer that while space did not permit a detailed description of all the work done on this subject, it would be worth while here to briefly refer to many of the reports made by American writers referring especially to those most easily available to members of A.F.A. and A.S.T.M.

Rother and Mazurie^{1†} made a series of test bars 1 in., 1½ in., 2 in., 2½ in. and 3 in., which showed the decrease in strength as the diameter increased. The percentage of such decrease depended somewhat on the mixture used.

A little later this question was referred by Committee A-3 of A.S.T.M. to a Sub-Committee on Correlation of Test Bar and Casting formed in 1926. The annual reports of this committee, beginning in 1928, should be read by all those interested. An extract from the report² of this Committee A-3 in 1928 reads as follows:

* Chief Chemist, Allis-Chalmers Mfg. Co.

† Reference numbers as shown herein correspond with Bibliography at end of the paper.

NOTE: This paper was the fourth of eight contributions presented at a Joint Meeting of the A.F.A. and the American Society for Testing Materials held as a part of the 1933 Convention of the American Foundrymen's Association.

"A number of the members of the committee have felt that one size of test bar will not give results that are representative of the various sections used in gray-iron castings. Consequently it was urged that the committee set up three distinct sizes of test bars to represent light, medium and heavy section castings."

Thus was forecast the present tentative specification which comprises bars of three diameters to represent various sections and which was proposed and adopted as tentative by the A.S.T.M. in June, 1932.

In our 1929 report³ an attempt was made by F. B. Coyle to express by a formula the relation of strength to section of casting. Articles on this subject appeared also in the A.S.T.M. Symposium on Cast Iron,⁴ in 1929. Mr. Rother here restates some of his results as follows:

"With increases up to about 2 in., the strength changes rapidly. Above 2 in. the variations in average structure are not so great, and the decrease of strength with increase in dimension is not so pronounced."

The writers' tests⁵ given at the same symposium, made on bars cast 1.25 to 4.0 inches in diameter, show a loss in tensile strength as the sections increase, about as follows:

- (1) For soft cast iron, silicon 2.25, nearly 8 per cent loss for each increase of 1 inch in thickness.
- (2) For a 25 per cent steel mixture, silicon 1.25, a loss of 5.7 per cent for each increase of 1 inch in thickness.
- (3) For a high-test iron, silicon 2.25 and nickel 1.4, a loss of 2.6 per cent for each 1 inch increase in thickness.

It should be noted that, especially in the first two grades of iron, the decrease was more rapid in the smaller sizes and the curves tended to flatten out as the bars increased in diameter.

In the Sub-Committee report of 1930, articles by A. L. Boegehold⁶ and M. V. Healy⁷ amplified the work already done along this line.

In the report⁸ of 1931, H. Bornstein reported upon a comparison of various sizes of test bars representing castings from five foundries. Jas. T. MacKenzie reported on Tests of Cast Iron Specimens of Various Diameters and J. W. Bolton on the Correlation of Test Bar and Casting by the Volume Surface Ratio.

The latter states briefly that test bars approximating the cooling rates of castings of simple shape may be chosen by calculating

the volume-surface area ratios of the castings and selection bars whose volume-surface area ratios are similar. In this way the speed of cooling is proportionally the same, and tests made of these test bars will closely represent the strength of the iron in the casting itself. This author is entitled to great credit for this novel and promising development, and the report will amply repay anyone for the time spent in looking it over.

The British point of view (in respect to separately cast test bars and several sizes of bars) is similar to our own, but the French prefer the shear test. In this test a small specimen is hollow drilled from the casting itself. Shear tests are then made on it and these results translated into their tensile values. In this way the casting may be explored to great advantage.

A paper⁹ on this subject was given in 1926 before the A.F.A. by G. K. Elliott, with discussion by W. H. Rother. Another paper¹⁰ was given before the A.S.T.M. in 1931, by H. H. Judson. In general, however, the American point of view is that, while the shear test is of value as a research tool, the transverse and tensile tests on separately cast bars are to be preferred for general use.

For many years the standard American test bar had been supposed to represent the iron in the ladle and not the iron in the casting. In 1932, however, Committee A-3 finally recommended to the A.S.T.M. as a tentative standard, Specification A48-32T which divides cast iron into seven different classes along the lines of their tensile strength. Test bars of three diameters are used to represent various castings, according to section size. Specifications and test bars will be discussed in some of the following papers at this session.

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Factors of Mechanism of Testing

BY JAS. T. MACKENZIE,* BIRMINGHAM, ALA.

In the testing of cast iron, there are points in relation to the equipment and its manipulation which are of special importance in securing accurate data. These are discussed briefly for the benefit of the less-experienced testing machine operators and those who contemplate testing cast iron.

Transverse Test.

Errors in load measurement in the screw machines are due to worn or dirty knife edges, looseness of the driving screw on the scale beam and ill adjustment of the balancing weights. The hydraulic machines are subject to bent pointers, friction and hysteresis losses, leakage, and adjustment of the lever weighing system in some types. The fact that a machine gives the correct reading in the static condition of a calibration does not prove that it will register the correct load in a rapidly made test. For that reason, tests made very slowly are generally more accurate than those made rapidly.

The measurement of deflection is one of considerable difficulty. All measurements dependent upon the movement of the head of the machine are apt to give errors of commercial magnitude. Cross-heads tilt, especially the heads of the large universal type machines. Yokes that are hung from the bar, as in the common machines built solely for transverse testing, are much more reliable.

A common source of error in such machines is in reading the cross-head movement after the bar has broken, neglecting the fact that it drops from 0.03 to 0.10 inch (and in some old machines more) *after* the bar is broken. This is due to the necessary play in the screws. The cross-head is pulling against the bottom of the threads at the break, and then sits on the top of them afterward. If the observer keeps his eye on the vernier and catches the reading at the break, fairly accurate results are obtained.

* Metallurgist, American Cast Iron Pipe Co.

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The ordinary string-over-a-pulley method of transmitting deflection to a drum is especially prone to error and requires practically frictionless bearings for success, as the string (or wire) stretches and then jumps too far, etc. It is almost impossible to get a decent measurement starting from the zero load point. If the attachment is good mechanically, a satisfactory reading usually can be made by starting the crosshead, with the connection to drum already made, at some distance above the bar. This gets the deflection device to moving before the load comes on, takes all of the slack out of the system, puts the string or wire in the proper tension, etc. It gives a very convenient diagram of the form shown in Fig. 1-A.

All the string-over-pulley devices known to the writer fail if attached to the bar itself, for the reason just stated.

The bottom of the bar is, however, the most accurate point to use for deflection. A lever-type extensometer with a factor of at least ten works well, and a dial gage reading to thousandths is convenient for direct reading if provision is made for protection of the gage from the slap of the broken bar. Care must be taken, especially in testing round bars, to see that the contact head of the gage is large enough to allow the bar a little lateral motion on the supports without changing the reading (illustrated in Fig. 1, *E* and *F*.)

A convenient form used by the writer for testing large speci-

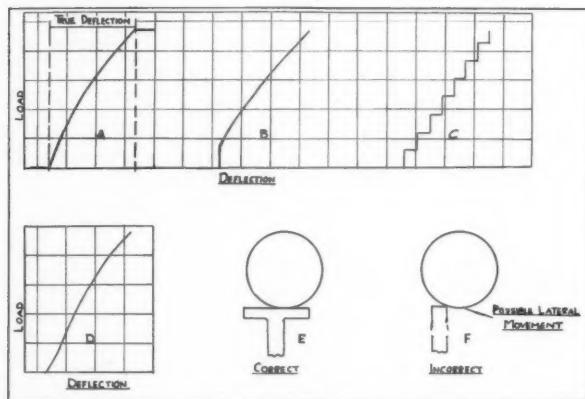


FIG. 1.

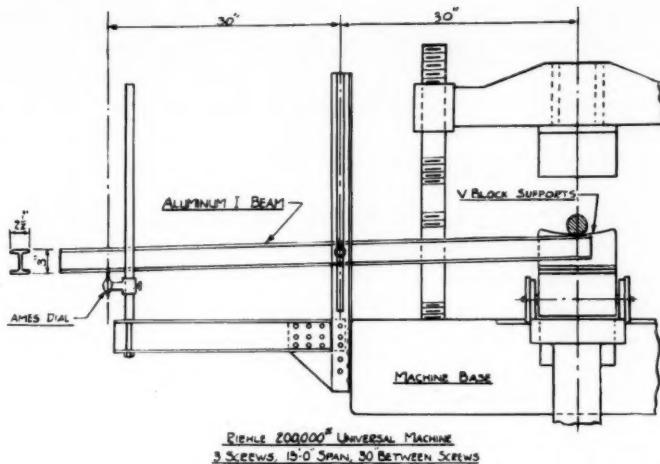


FIG. 2—DEFLECTION MEASURING DEVICE ATTACHED TO 200,000 LB. RIEHLE MACHINE.

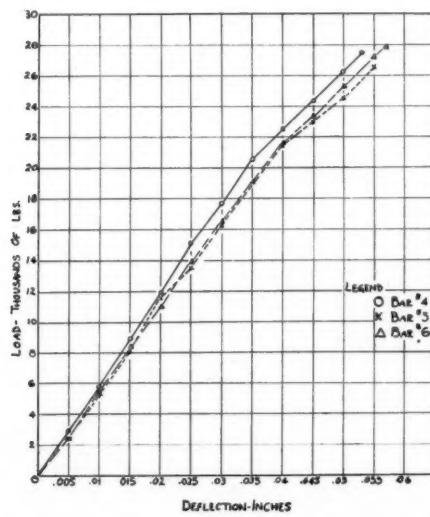


FIG. 3—LOAD DEFLECTION CURVES, 2.20-INCH DIAMETER, 8-INCH SPAN.

mens in a big machine is shown in Fig. 2. This consists of an aluminum I-beam under the specimen and a dial gage following the other end of the beam at an equal distance from the fulcrum on the edge of the bed plate. At the breaking point the beam jumps up several inches but is caught by the band before it falls back on the gage. A typical set of curves (three bars of the same iron) on 2.20-inch diameter and 8-inch span—a very small total deflection—is shown in Fig. 3. This test procedure requires three men; one to balance the beam, one to read deflection, and the third to record the readings.

Whatever type is used, it is imperative to plot the results and examine the curves carefully to see if the deflection proceeds smoothly with the load. Several types of curve resulting from error in deflection or load readings are shown in Fig. 1, *B*, *C* and *D*.

Tensile Test.

The tensile test is, in addition to errors in the machine weighing system, quite sensitive to errors in alignment of the specimen as a transverse stress reduces the tensile values rapidly. For the highest accuracy some such device as the Robertson shackles should be used.

Cast iron is not particularly sensitive to surface finish, neither are fillets so important; but it is well worth while to prepare the specimen exactly according to drawing and give the surface at least a fair polish.

Hardness Test.

Hardness is best determined by the Brinell test, using a ball of 10 mm. diameter with a load of 3000 kg. Good judgment is necessary in selecting a sufficient number of points to give a complete picture of the casting. The machine should be calibrated from time to time and the ball carefully watched and discarded at the first sign of flattening. No reading should be accepted which shows a crack—for instance, in an impression made too close to the edge—as such a reading will be low. In testing thin or small castings the 5-mm. ball and a 750-kg. load is useful, as is also the Rockwell machine with the 1/16-inch ball and the 100-kg. load (B scale). Any results from these small balls should be taken only if sufficient readings have been made to establish definitely the hardness as, especially in soft irons, they are likely to give very erratic results.

Factors in the Production of Test Bars

BY JAS. T. MACKENZIE,* BIRMINGHAM, ALA.

The rules for molding and pouring test bars are the same as for any other casting of similar shape and section. The industry suffers in a great many instances from the indifference of the molder on this job who expresses himself by the statement, "It's just a test bar."

The results on bars tested in the "as cast" condition are sharply influenced by the condition of the surface, the high-test cast irons being especially sensitive to surface conditions while the soft irons are not. Particularly troublesome are the dirt or slag particles that are caught just in or partly beneath the skin of the bar, and this trouble is by no means confined to bars cast on the side.

A few of the most important desiderata in the molding and pouring of test bars are as follows:

- (a) A strong, permeable and refractory sand.
- (b) A facing which adheres well to the sand and peels cleanly.
- (c) Gates placed so that the middle or highly stressed part of the bar is not subject to shrinkage or uneven heating, and does not have to be chipped or ground.
- (d) Molds should have a good finish without slicking or swabbing. Fins which produce chilled spots on the bar are especially obnoxious.
- (e) Pouring should be neither too hot nor too cold. In general, a temperature should be used which will just fill the mold with liquid iron. This will be about 100 degs. Cent. above freezing point for a 2.0 inch bar, 150 degs. Cent. for the 1.20 inch bar, and 200 degs. Cent. for the 0.875 inch bar. At such temperatures, reasonably slow pouring gives the best results in giving clean, sound bars. If the iron must be poured colder it should be poured faster, and *vice versa*.

* Metallurgist, American Cast Iron Pipe Co.

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- (f) A thoroughly dry ladle should be used, preferably pre-heated to at least a dull red or higher.
- (g) A skim gate should always be used. If not, very careful skimming is necessary.
- (h) A runout mold should never be "caught up" as this is, in effect, then a hot mold.
- (i) In high-carbon irons, horizontally cast bars often give considerably higher carbon and more open grain on the cope side. A vertically cast test bar is preferable for such material.

Wear Tests and Value of Hardness Testing for Control of Product

A. L. BOEGEHOOLD,* DETROIT

The first section of this paper is devoted to a description of various types of wear tests which have been conducted by investigators in the past. In all of the reports of wear testing examined, it was found that no two were conducted under like conditions. No investigator apparently has ever attempted to check the work of another. However, in spite of this lack of agreement in making tests to determine resistance to wear, it is remarkable that so many of the conclusions agree.

Methods for conducting wear tests may be classified in several groups, somewhat as follows:

- (1) Service tests.
- (2) Accelerated service tests.
- (3) Laboratory tests in which service conditions are simulated as nearly as possible.
- (4) Laboratory tests under some arbitrary set of conditions.

CLASS 1—SERVICE TESTS

A large amount of wear testing of this type is continually going on. Such tests are the most conclusive of any wear tests and also are probably the most expensive. By far the greater part of the results obtained with this type of test are never reported in the literature.

A few examples of testing in this class which have been reported are the work of Kuhnel¹† who tested various grades of cast iron in locomotive valve rings and valve boxes. A German committee of iron founders and railroad men² conducted tests of cast iron in brake blocks in actual service on railroad carriages. Gallwitz³ made tests of materials for plowshares; his service tests made on plows in the field were made for the purpose of correlating with a larger number of laboratory tests.

* Metallurgist, General Motors Research Laboratories.

† Reference numbers as shown herein correspond with the Bibliography at the end of the paper.

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CLASS 2—ACCELERATED SERVICE TESTS

This type of testing is represented by the work of Bornstein,⁴ who tested the wear in tractor engine cylinders by running the tractor for 20 hours in high gear under the following conditions:

- (1) Crankease oil—Mobile BB diluted one-third with distillate.
- (2) Filter collar removed from air cleaner.
- (3) Dirt or abrasive injected at rate of $\frac{1}{2}$ lb. per hour for 6 hours.

CLASS 3—LABORATORY TESTS SIMULATING SERVICE CONDITIONS

Results obtained in laboratory tests can be relied upon only when materials of known value in service are first subjected to the laboratory test and are rated in the same order as determined in service. When this can be done, the laboratory test becomes of great value because it permits the testing of a much greater variety of materials than can economically or conveniently be tested in service. The attainment of the test conditions which will give this result is sometimes very difficult. The logical course to pursue, obviously, is to make the test conditions as near like service conditions as is possible.

Examples of tests which have aimed in this direction are given below.

Saniter⁵ simulated the action of rolling friction encountered on railroads by rotating a horizontal cylindrical specimen which drives the inner race of a ball bearing producing a rolling motion between the specimen and the inner race of the ball bearing. The inner race diameter of the ball bearing is considerably larger than the diameter of the specimens. The Saniter wear number is based on the reduction in diameter of the test piece in ten thousandths of an inch during a certain number of revolutions of the test piece compared to the loss of a standard bar under similar conditions. The Saniter test as conducted by the National Physical Laboratory⁶ consisted of a cylindrical specimen 1 inch in diameter revolving at 2200 r.p.m. A hardened steel ring $1\frac{1}{2}$ inch inside diameter rolls on the 1-inch specimen with the inside of the ring in contact. This hardened ring is surrounded by a ball bearing mounted in an adapter from which is hung a load of 410 lbs.

Lucchini and Ros⁷ also were interested in wear of rails. They used an Amsler machine with test pieces in the form of 4 cm. diameter discs, 1 cm. wide, pressed together under 40 kg. and

rotating at speeds of 200 and 180 r.p.m., respectively, in order to produce slippage. The slippage is somewhat greater than experienced between locomotive wheels and rails, but results were stated to agree with service results.

Hengstenberg⁸ studied the resistance of metals to erosion by water impact. Two specimens mounted oppositely on an impeller cross the path of jets of water from two nozzles parallel to the impeller shaft. The erosion produced in a few minutes at a peripheral speed of 600 to 1200 ft. per second represents several years of average operating conditions.

Rogers⁹ studied the wear resistance of steels in contact with moving abrasives. Two specimens $\frac{1}{2}$ in. in diameter and $\frac{3}{16}$ inch thick are carried at the ends of arms revolving inside a casing which has fixed ribs placed to prevent swirl of the mixture of water and coarse emery with which it is partly filled. The loss of weight is measured after running for a period of 10 minutes at 1500 r.p.m.

Gallwitz¹⁰ arranged a laboratory test simulating the conditions causing wear of plowshares. The shares were mounted on a revolving frame passing through sand held in a ring-shaped cement canal at 3 meters per second. The cross-section of the sand ring was 80 cm. wide by 40 cm. deep. Building sand from 1 to $\frac{1}{4}$ mm. grain size (mostly $\frac{1}{2}$ mm.) was used kept at 10 per cent moisture. Weight losses were 5 to 30 g. per 10-km. run in the laboratory, and 30 to 600 g. in field tests per 3-km. run. Good correlation was obtained between laboratory and field results.

Blake¹¹ used the Fahrenwald machine to test steels under conditions encountered in wet grinding operations. Test specimens under load were moved in contact with abrasive material and water in an annular trough with controlled abrasives, speed, pressure, time and moisture.

Hall¹¹ used a small stone crusher fitted with jaws made of the steel to be tested in determining the best steel to resist the action of stone crushing. He noted that Amsler and Brinell machines did not give results that indicated how a steel would behave in service.

Parker¹² tested tire chains by rubbing them on a grinding wheel under controlled conditions.

French and Herschman¹³ built a machine which simulated the conditions attending the inspection of holes in metals by means of plug gages.

CLASS 4—LABORATORY WEAR TESTS UNDER ARBITRARY CONDITIONS

Robin¹⁴ rubbed test specimens against a rotating disc faced with emery paper. Loss of weight was measured after rubbing 200 meters. Diameter of the rubbing track was 150 mm. The sample was 15 mm. in diameter. The rubbing speed was 1.15 meters per second and loaded with 1 kg. per sq. cm.

The Brinell abrasion testing machine is described by Rosenberg and Herschman.¹⁵ This machine produces wear by the action of sand fed between the specimen and the periphery of a disc of open-hearth iron pressed against the specimen. Rosenberg¹⁶ used this apparatus to test the resistance of steels to abrasion by sand.

A number of investigators have used the brake shoe test in which the specimen is stationary and is pressed against a rotating steel or cast iron drum in a manner similar to the brake shoe on a car wheel. This test has been used by Lehman,¹⁷ Boegehold,¹⁸ Scharffenberg,¹⁹ Lowry²⁰ Spazier²¹ and others. The Spindel machine used by Piwowarsky²² and Klingenstein²³ is a modification of the brake shoe test in that the rotating drum is so thin that it becomes a disc. The test specimen pressed against the thin edge of this rotating disc is cut as if by a grinding wheel. The length of this incision or the depth after testing for an arbitrary length of time is the measure of wear resistance.

Another form of drum test was used by Swift²⁴ in which the flat side of the drum was used as the rubbing surface for three test specimens evenly spaced around the track 120 degrees apart. This type of test lends itself well to the determination of frictional coefficient.

The test conditions used for the above drum type tests are given in Table 1. It will be noted that in all cases the rubbing speeds are comparatively low and, except for tests by Spazier, the loads are quite low. These are obviously not the conditions that would ordinarily exist in service. So far as the work done by the author is concerned, it was found that at the rubbing speed used, the load could not be increased very much without danger of galling and scoring occurring, especially in some of the low-combined-carbon irons. This consideration presumably influenced the other investigators to use low loads and speeds for their test conditions. This may be cited as one objection to the practice of conducting wear tests without lubrication.

Swift, in common with the author and others, in discussing

Table 1
ROTATING DRUM.

Investigator.	Drum Dimensions—Width, Inches.	Point of Contact with Specimen	Rubbing Speed, ft. per min.	Pressure lbs. per sq. in.	Specimen Tested, inches	Duration of Test	Loss of Weight.	Lubrication.
Lehman	4 $\frac{3}{8}$	1 $\frac{1}{8}$	Periphery	457	22 114	1.180 985x0.79	2 hrs.	0.43 to 26.1 gm.
Boegehold	8	2 $\frac{1}{2}$	Periphery	1600	12	0.46 Comb. Carbon Cylinder Iron 180 Brinell	24 hrs.	0.50 to 6.00 gm.
Swift	5.35 Mean Diam. of Rubbing	Flat Side	112	10, 20, 30	Nitralloy Gear Steel, M11d Steel, Cast Iron, Phos. Iron, Bronze, Halo Brake Lining.	0.57 x 0.44 x 0.1875 Thick; 3 Specimens	100 hrs.	none
Robin	5.7 Mean Diam. of Rubbing Track	Flat Side	226	14.2	Emery Paper	0.59 diam.	2.9 min. (rubbing to 200 meters)	Milligrams per sq. cm.
Piwowarsky	11.8 (300 mm.) x .039 (1 mm.) Spindle Machine	Periphery	77.3	Starts High and Decreases as Cut Lengthens. Load 6, 11, 22 lbs.	0.23% Carbon Steel	10 min. Wear by Length of Cut Made.	none	none
Lowry	Not given.	Periphery	77.3	(Insufficient Data) 630 r.p.m.	15 lbs. wt. at End of 3 ft. Lever.	1000 hrs.	none	none
Spanier	1.575 (4 cm.) x 390 (1 cm.) Amster Type Machine	Periphery	77.5	12,000 lbs. Load	Steel	1.18 x 0.39 3 cm. x 1 cm.	0.0015 to 0.03 gm. per sq. cm. of contact surfaces.	none
Klingenstein	3.94 (10 cm.) x 195 (5 mm.) Spindle Type Machine	Periphery	415	11 lbs. Load	Steel	3 $\frac{1}{4}$ min. (1500 rev.) cut determines. wear.	Depth of groove (1500 rev.) cut determines. wear.	none

Table 2
AMSLER TYPE WEAR TEST.

Investigator.	Diameter of Rollers, Inches.	Width of Rollers, Inches.	R.P.M. of Rollers.	Rubbing Speed, ft. per min.	Pressure at Rubbing Surface, lbs. per sq. in.	Duration of Test.	Loss of Weight.	Lubrication.
Piwowarsky	1.18-1.97 (30-50 mm.)	0.394 (1 cm.)	No. 1-0 No. 2-200	103	50 kg. Load	60 min.	0.004 to 0.017 gm.	yes
Luchini & Ros.	1.575 (40 mm.) Both Rollers	0.394 (1 cm.)	No. 1-300 No. 2-180	8.25	40 kg. Load			no

the reasons for conducting tests without lubrication to find out how materials will wear when lubricated, says that if lubrication is perfect, no wear occurs at all; if it is imperfect, it is next to impossible to control, and wearing action depends as much on intermittency of lubrication as on any intrinsic property of surfaces in contact.

We should realize that in actual service the rubbing speed and the unit pressure prevailing at the time when metal-to-metal contact is established between two imperfectly rubbing surfaces, is usually greatly in excess of the speeds and pressures adopted for laboratory tests without lubrication. In spite of this discrepancy, the results obtained by testing dry seem to indicate what service results will be.

Pirowarsky²² also used the Amsler machine, keeping one disc stationary. This makes the test similar to the brake shoe test, except that the contact area is constantly changing and the unit pressure continually diminishing from a very high value when the stationary disc is still completely round, down to a much lower pressure after wear has taken place. The loss of weight in the presence of lubrication being quite small, the change in unit pressure would not be appreciable. The conditions used for this test are given in Table 2.

Klingenstein²⁵ used a test which is somewhat of a departure from commonly used wear tests. His cast iron specimen was in the form of a blunt drill 1 cm. in diameter which rotated at 325 r.p.m. while pressed against a flat cast iron piece 3 x 3 x 1 cm. under a pressure of 24 kg. per sq. cm. (598.5 lbs. per sq. in.) The sum of the weight losses of the flat and the cylindrical specimens after ten minutes running without lubrication was the measure for wear resistance.

CONCLUSIONS CONCERNING WEAR OF CAST IRON

With a number of investigations, each year reporting their results on the wear testing of cast iron, it should soon be possible to form a very definite idea about the wear resistance of this material. How much better it would be if two or three types of wear tests for studying wear under different circumstances could be selected and a standardized test procedure established for each type of test so that the results of each investigator's works could be correlated with every other work on the same phase of the subject.

A summary of the conclusions reached by the various investigators who have tested the wear resistance of cast iron will give an idea as to the value of the various types of tests.

Lehman,¹⁷ using the brake shoe type test, concluded that:

- (a) Brinell hardness does not indicate the frictional resistance quality.
- (b) Chemical composition cannot be used as a guide for judging abrasion qualities.
- (c) Abrasion resistance against steel is increased with increasing pearlite content and is influenced unfavorably by the presence of phosphide eutectic.
- (d) Gray cast iron with full pearlitic structure has the minimum abrasion when run against steel or cast iron.

Scharffenberg,¹⁹ using the brake shoe type test:

- (a) Brinell hardness and wear showed a certain relation.
- (b) Silicon increases wear at contents of more than 1.7 per cent.
- (c) Manganese decreases wear slightly up to 1.5 per cent Mn, above which content the wear increases greatly.
- (d) Phosphorus and sulphur both cause diminished wear as these elements increase.
- (e) Temperature occurring during the wearing does not indicate the resistance to wear.

Klingenstein,²³ using a brake shoe test of the Spindel type:

- (a) In annealed cast iron (ferritic) below 160 Brinell hardness, the wear showed little change up to 0.55 per cent phosphorus; from 0.55 to 1.9 per cent phosphorus, the wear was reduced 33 per cent.
- (b) In pearlitic cast iron not annealed, having Brinell hardness of 210 to 240, wear was very little affected by phosphorus up to 0.70 per cent. From 0.70 to 1.35 per cent phosphorus, the wear was reduced 25 per cent, with no further effect above 1.35 per cent phosphorus.
- (c) The pearlitic iron of 210 to 240 Brinell hardness wore only one-half as much as the annealed iron of 160 Brinell.

Klingenstein,²⁵ using his blunt drill test, found that:

- (a) Minimum wear occurs when the two parts are of equal hardness.

- (b) Phosphorus increase from 0.4 to 1.0 per cent, decreased wear 42 per cent.
- (c) Chromium increase from 0.3 to 0.7 per cent, decreased wear 25 per cent.
- (d) Chromium plus nickel increase from 0 to 1.8 per cent, decreased wear 25 per cent.
- (e) Manganese increase from 1.0 to 1.65 per cent, decreased wear 8 per cent.
- (f) Silicon increase from 1.15 to 2.0 per cent, increased the wear 65 per cent.

Piwowarsky,²² using the Spindel type test dry and the modified brake shoe test on an Amsler machine lubricated, found that resistance to wear is greatly increased with increasing phosphorus up to 0.8 per cent.

Bornstein,⁴ in an accelerated service test, found that tractor engine cylinders containing 0.70 per cent nickel and 0.22 per cent chromium, having a hardness of 143 to 159 Brinell, wore 25 per cent less than cylinders without alloy having a hardness of 118 to 130 Brinell.

Boegehold,¹⁸ using a brake shoe test, found that:

- (a) Wear resistance of cast iron in contact with cast iron containing 0.46 per cent combined carbon increases with combined carbon, reaching a maximum at 0.80 per cent combined carbon.
- (b) Permanent mold cast iron wore at a rate many times faster than sand-cast iron of the same combined carbon content.

Swift,²⁴ using a drum test with specimen rubbing on the flat side, concluded that:

- (a) As a rule, a metal wears much more rapidly against itself than against any other metal, and in a general way the more different the nature of the two metals, the better do they resist wear.
- (b) Other conditions being equal, it is expected that a hard material will resist abrasion better than a soft one, and in general this is confirmed by the present tests.
- (c) Suitability of materials for wear should be determined not as individuals but as pairs.

Lowry,²⁰ using a brake shoe test, found that wear decreased progressively with increased phosphorus from 3.75 per cent for 0.195 phosphorus to 1.5 per cent for 0.911 phosphorus.

He gives a chart purporting to show no relation between Brinell hardness and wear. It is difficult to interpret this chart, however, because points are recorded representing irons which apparently have Brinell hardnesses of 42, 44, 50, 58, 65, 80 and 90 Brinell, as well as others between 110 to 220.

He mentions the glazing of soft cast iron when running against hard cast iron so that the soft iron wore less than the hard iron. More rapid wear occurred when two hard irons were run together. This does not agree with Klingenstein's results, which showed the least wear when two irons of the same hardness were run together.

Robin,¹⁴ using an abrasion test with emery paper, found that:

- (a) Resistance to wear increases with the amount of free cementite and with the percentage of phosphorus.
- (b) Free graphite does not appear to influence the results.

Spazier,²¹ using a brake shoe type test on an Amsler machine, found that:

- (a) The lowest rate of wear occurs with high combined carbon and high phosphorus.
- (b) Wear is not related to tensile strength.
- (c) Specimens over 210 Brinell had much lower wear than softer ones.
- (d) A specimen of 200 Brinell wore about six times as much as one which was 169 Brinell.

Kuhnel¹ found in testing locomotive valve rings and valve boxes that ring failure was greatest below 140 Brinell, and that with a 10 point difference between the hardness of ring and box, the wear is least, being only half that when both are of the same hardness. The greater the difference in hardness between ring and box, the greater the wear.

A German committee of iron founders and railroad men² found that, in general, wear decreases with increasing hardness, but with many exceptions. Best results were obtained with pearlitic structure.

General Conclusions

Certain conclusions to the contrary, the general picture obtained from all this work is that increased hardness, or the addi-

tion of elements which form hard constituents in the iron either by going into solution or by combining chemically, results in decreased wear. It is unfortunate that those who have published conclusions contrary to this picture have not given more details as to the microstructure and history of the metals tested or other details which might explain the deviation from a naturally expected result. The reason for exceptions to the general conclusions above is not now known, so that future work on wear testing should be conducted for the purpose of establishing these causes.

VALUE OF HARDNESS TESTING FOR CONTROL OF PRODUCT

Hardness testing has been and will continue to be a satisfactory means of inspecting the quality of cast iron articles. The value of hardness testing for control of product does not depend upon any broad relationship between hardness and other physical properties of cast iron. By selection of the proper examples from various kinds of cast iron, it can be shown that hardness has no relation whatever to strength, machinability or resistance to wear. On the other hand, it can be shown that the normal variation in hardness experienced during the daily manufacture of one grade of cast iron, is directly related to strength, machinability and wear.

Having established by service tests the strength, wear resistance or other physical property limits necessary in the cast iron for a particular part, hardness tests can be made upon irons possessing properties at the upper and lower limits. These values for hardness then become an empirical yardstick which may be used as a convenient means for daily or hourly checking the product. Having established the relationship between hardness and the physical properties necessary to the successful performance of the casting, the hardness test will tell whether the product falls within the specified property limits, just as well as will tests for the properties themselves.

Physical property tests such as resistance to wear, tensile or compressive strength, etc., are much more troublesome to make than a hardness test, consequently a great convenience is afforded by substituting the hardness test. The advantage of the hardness test is that for important castings, every casting can be tested, thus providing assurance that each casting shipped will give satisfactory performance. This, obviously, would be impossible with physical property tests, which usually require destruction of the casting.

Just as physical properties of castings are calibrated in terms of hardness in order to get a control test to be used for rejection or acceptance of castings, another calibration may be made in terms of depth of chill obtained when the iron is cast against a chill block. This test is frequently used for testing the molten metal being poured into castings. Such a test can be made long before a hardness test could be made on one of the castings, consequently, any deviation of the iron from the quality desired is quickly realized and adjusted before too many castings are poured from "off" iron.

Many examples could be given where hardness is used as an acceptance test. A few will illustrate the point. Piston rings are specified between Rockwell B 98 and 106. Cylinder iron for certain passenger cars must be between 163 and 207 Brinell hardness. Below the lower limit, excessive cylinder wear and valve seat wear may occur; above 207, machining becomes too difficult for the speeds and feeds which have been set for this particular grade of iron.

Cylinder iron for certain truck or bus engines must be between 207 and 241 Brinell hardness. Brake drum iron must be between 228 and 269 Brinell hardness. Below 228 Brinell, wear resistance is insufficient and strength falls below the required 45,000 lbs. per sq. in. minimum. Above 269 Brinell, machining trouble is encountered. Obviously, on parts of this kind, a hardness test is about the only kind of a non-destructive test that can be made which can be translated in terms of physical properties.

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A.S.T.M. Specification A48-32T

BY JOHN W. BOLTON,* CINCINNATI

The A.S.T.M. specification for gray iron castings (Specification A48-32T) is in some respects radically different from the specifications for gray iron castings and for high-test gray iron castings which preceded it. To clearly understand the significance of the present specifications it is necessary to review briefly the various factors which have influenced its evolution.

HISTORY OF A48-32T SPECIFICATIONS

The principle on which the older specifications were based was that a cast-iron test bar could be expected to indicate little more than that hypothetical measure, the so-called "quality of the iron in the ladle." While the early work of Keep, West, Moldenke and others had clearly demonstrated that properties of cast iron varied with changes in sections, little was known about the quantitative influences of the various factors which caused these variations, and practically nothing was known about methods for correlating properties of test bars and castings.

The so-called arbitration test bar has been in use since 1905 and it was recognized that most foundrymen and engineers were unfamiliar with the radical changes proposed. Therefore, a campaign of education was undertaken informally by the committee.

The report of Committee A-3 for 1928 contains the following significant paragraph: "The arbitration test bar for the transverse testing of cast iron has received further consideration during the past year. A number of the members of the committee have felt that one size of test bar will not give results which are representative of the various sections used in gray iron castings. Consequently, it was urged that the committee set up three distinct sizes of test bars to represent light, medium and heavy section castings."

It was felt that the data then available were not sufficient to allow formulation of a comprehensive specification. To meet this objection the A.S.T.M. encouraged a number of investigations, including those of J. L. Jones, R. S. MacPherran, A. L. Boegehold, M. V. Healey, H. Bornstein, J. T. MacKenzie and others. In 1929

* Metallurgist, The Lunkenermer Co.

NOTE: This paper was the eighth and concluding contribution presented at a Joint Meeting of the A.F.A. and the American Society for Melting held as a part of the 1933 Convention of the American Foundrymen's Association.

it was recommended that a closer and more definite classification of irons by classes based on ranges of tensile strength be drawn up by Committee A-3, such classes to form a more definite basis for a suitable specification.

In June 1931, Committee A-3 passed the following resolution:

(1) It is felt that the engineering fraternity needs more information and assurance as to the quality of the iron in the castings than is given by the present specifications which indicate the so-called quality of iron in the ladle.

(a) The present single test bar is inadequate in that it represents only one set section size. The properties of gray iron are considerably modified by change in cooling rate, which in turn is affected by section size.

(b) Therefore, it is deemed advisable to consider adoption of a series of three or more bars to cover a range of sections approximating the ranges found in usual commercial practice.

(2) The function of the specifications is to indicate the intrinsic quality of material possible for a given section or range of sections, and not to enter into the field of workmanship and design. It is recognized that no specification can assure the quality of a casting and entirely remove the responsibility resting with the manufacturers.

(a) It is felt that the committee can devise a set of specifications to better indicate the properties of iron in given sections, this to be done by means of physical test specifications for a series of test bars.

(b) This should comply with the policy stated at the March, 1929 meeting, as follows: "Cast iron shall be known and listed by classes wherever there is no application name now; for example, pipe iron, cylinder iron, etc. Any names to describe quality shall be abandoned; for example, high-test cast iron, semi-steel, etc."

(c) For each given size bar the classes are to be listed according to minimum tensile strength, as follows:

1—15,000
2—20,000
3—25,000
4—30,000
5—35,000
6—40,000
7—50,000

Further provisions are to be made for corresponding transverse tests and other test information.

(3) One of the main functions of this committee is to educate those interested in cast iron to the possibilities and limitations of specifications as applied to this material. It will be necessary to do this prior to actual publication of the newer specifications, since these will represent a radical departure from established conventional practice.

There was considerable discussion at this meeting, particularly relative to the series of test bars to be used. The 1.2 inch test bar was recognized as the tensile bar, but complete agreement was not reached on the smaller and larger sizes. Some time later a series of test bars was proposed by the Gray Iron Institute. This series as well as the British series of test bars was studied by the committee. Active cooperation of the various interested parties resulted in the development of tentative specification A48-32T, which was presented to the American Society for Testing Materials at the June, 1932 meeting.

It is recognized that this specification is far from perfection, and Committee A-3 welcomes suggestions which would help to improve this specification. It is known that considerable work must still be done on the subject of test bars and correlation on sizes of test bars in order to secure adequate data to improve the specification.

FEATURES OF THE SPECIFICATION

(1) *Classification by Tensile Strength.*

(a) The classification covers all the widely used commercial irons. In the specification, minimum tensile only is specified. It is not implied or intended that if an iron furnished as 25,000 minimum runs, say, 31,000 that the specification has not been met. The upper limit in strength is a matter up to the individual consumer and producer, and in large degree depends on the degree of machinability acceptable to the consumer. Obviously, if the producer furnishes an iron which has in any degree a mottled or chilled fracture, he no longer is furnishing truly gray iron castings and the consumer has recourse within the specification.

(b) Special attention is directed to Explanatory Notes 1, 2 and 3.

(c) Careful researches by various metallurgists have shown that if tensile tests are conducted with reasonable care, the tests

are accurate, reliable and comparable. Practically all engineering formulas use tensile strength as a factor in computations, and this fact has influenced the selection of the tensile test as a basis of classification.

(2) *Transverse Test.*

(a) Transverse tests are optional. The committee recognizes that there is no definite ratio between tensile and transverse tests. The values for transverse load given are, generally speaking, somewhat lower than the average that would be obtained from irons of the respective tensile groups.

(b) The committee has not yet determined what the most logical and equitable deflection values should be. There are many factors to consider, divergent points of view to be reconciled and further data to be obtained. It is possible that intermediate as well as ultimate deflection values may be specified.

(3) *Test Bars.*

The test bar sizes will represent in a fair way the effect of mass on cast iron. The present sizes were chosen after careful consideration of several series proposed, including the British series which has been in use for a few years. A properly chosen series of test bars not only should give some indication of the influence of mass on iron, but also some consideration should be given to providing a series which permits logical casting section correlation to be evolved. The data at hand relative to this problem are meager and are confined to the work of a few individuals. More research is needed. Test bars should be as few in number as is consistent with the actual needs of the trade.

The fact that the British employ three bars and that their specification has been established for several years, suggests that in the interests of international standardization it might be desirable to conform to their series. The two smaller bars in the A48-32T specification are identical in diameter and length with the British bars and are tested at the same span. Our largest bar is 2 inches in diameter against 2.2 inches for the British, and is 27 inches long as against 21 inches for the British. This permits our bar to be tested at a span of 24 inches instead of 18 inches. After careful consideration and discussion, Committee A-3 adopted the 2.0 inch bar for our largest size bar.

There is some demand on the part of manufacturers of light castings for a smaller bar than the 0.875 inch test bar, and some

work is now being done to determine the desirability of a smaller bar and the size of such bar.

CHEMICAL COMPOSITION

It is well known that irons of similar physical properties may be produced with a wide range of chemical analysis. Usually the consumer is buying castings for mechanical service and is interested in getting sound and sufficiently strong castings at the lowest over-all cost. Therefore, analysis or composition can be left to the foundryman. If other properties are required, such as corrosion resistance, wear resistance, machinability, etc., it may be desirable to hold certain elements within definite limits, and this is provided in the specification.

CERTIFICATION

Certification is a new feature in this specification. In a foundry where tests are made regularly, it permits the manufacturer to certify that his product conforms to the requirements of the specification.

COOPERATION

The present specification is an attempt on the part of producers and consumers on Committee A-3 to evolve a sound and equitable specification. It is a forward step in the effort to bring gray iron into the foreground as a leading material of construction. The needs of engineers and designers have been recognized. The major principles underlying specification are sound. More research and study are required to clarify some of the details and improve the specification. The writer urges the cooperation of all interested parties in this development.

WRITTEN DISCUSSION

W. WORLEY KERLIN:¹ The adoption of specification A48-32T by the American Society for Testing Materials is the most important step toward the recognition of gray iron as a reliable structural material in several decades. This belief is based on two facts which have been elaborated in the papers of this symposium: (1) Gray iron is not a material but a family of irons capable of classification; and (2) gray iron should be tested in bars which approximate casting sections.

Since the engineer believes that the tensile test gives more reliable design information and the gray iron founder uses the transverse bar as a control test, a careful correlation test should be made for each

¹ Gray Iron Founders' Association, Cleveland.

iron investigated by a particular founder. Information on transverse tests should include the following items: (1) A rigidity test; (2) transverse breaking load; and (3) transverse breaking deflection.

The last two tests are familiar to all. The first item consists in noting the deflection when a given load, less than the breaking load, is applied to the bar. The amount of that load is governed by the tensile class to be met. A load that will give a maximum fiber stress, of $\frac{3}{4}$ of the tensile strength, as calculated by the beam formula, has been found to be most serviceable. For example, a Class 20 iron should be stressed to 15,000 lbs. per sq. in.; a Class 30 iron to 22,500 lbs. per sq. in., etc. This deflection should be corrected to standard diameters and a maximum allowable deflection found for each type of iron.

In applying these tests to foundry control, the rigidity test generally will be the first to fail if the iron in a series of runs becomes weaker. That indicates the danger point at which measures should be taken to prevent further reductions in strength. By this method, it is possible to keep an iron above the minimum requirements of a given A.S.T.M. number. The Gray Iron Founders' Association has records of foundries which have done this 100 per cent for every heat since last August with an average deviation from the mean of 5.3 per cent.

Maximum physical requirements often are as important as minimum limits because they indicate the point where machining difficulties of a higher class number may be expected. In this case, the transverse breaking load seems to indicate the first departure from average practice. This should be the danger signal where measures should be taken to prevent hard castings. This criterion may be used to compare an iron of a given type over a series of runs. However, it cannot be applied in comparing one type of iron with another produced under a different set of conditions.

It often has been said that gray iron castings are seldom sold on a purely tensile strength basis and, while this is true to a certain extent, the thought should not be dismissed without closer examination. Many of the most desired properties such as close grain in heavy sections, machinability in light sections, toughness, rigidity and others may be associated with certain tensile classes. Other pertinent data may supplement a given tensile classification to advantage once this fundamental point of agreement is reached between founder and engineer.

ORAL DISCUSSION

J. T. MACKENZIE²: I would like to discuss the transverse test because of the statement which is made so generally that the tensile strength is what the designing engineer wants. I am connected with the cast iron pipe industry and one of the principal items with which we have to contend is transverse loading. A cast pipe of the old general character that we have used and which we still have to use in sizes up to 48 and 60 inches will show a tensile strength of 13,000 lbs. per sq. in. as calculated by the pipe formula from a destructive test. If

² American Cast Iron Pipe Co., Birmingham, Ala.

the ring of that pipe is loaded in a testing machine, with earth or in any other manner, the modulus of rupture will be approximately 40,000 lbs. per sq. in.

Why design cast iron pipe using a fiber stress of 13,000 lbs. per sq. in. to calculate transverse resistance when the figure that should be used is 40,000 lbs. per sq. in.? There is no justification for such calculations.

Take the case of the scale manufacturer. He is interested in a big, heavy beam. The grade of cast iron in which he is interested has a tensile strength of approximately 15,000 lbs. per sq. in. for a scale beam 10 in. deep by 8 in. wide or 16 in. deep and 4 or 5 in. wide but his useful modulus of rupture, if he is using a fairly long span, will be 60,000 lbs. per sq. in. Since most scale makers use short spans, the modulus probably will run 70,000 lbs. per sq. in. The modulus of rupture increases as the span decreases.

If the scale manufacturer makes a transverse test, he can know how much his beam will bend. In a tensile test, unless he has an accurate extensometer and uses it carefully, he does not obtain results which will permit him to calculate the modulus of elasticity or draw a stress-strain diagram. If he makes a transverse test, he knows the strength and deflection and can calculate the modulus of elasticity easily.

Here is the information he wants. He will use the 70,000 lbs. per sq. in. modulus of rupture and give himself a safety factor of $3\frac{1}{2}$ to 1, which is plenty. That gives him a fiber stress of 20,000 lbs. per sq. in. From the transverse test, he can use the modulus of elasticity of 15,000,000 lbs. per sq. in., figure his deflection and see what he has.

The French long have recognized this and also the difficulty of measuring the deflection at low loads. With the small specimen they use, they measure the deflection between 100 and 400 kilograms. Breaking strength of French test-pieces is approximately 500 kilograms.

When a good, low phosphorus, low carbon, well treated cast iron is tested, a different type of curve results. If you are as interested in cast iron as you are in steel, you will want to know the modulus of elasticity for the stiffness of your beam. What you want to know is, if the beam receives a certain blow will it or will it not break? It would be justifiable to set a maximum deflection at the useful load and still set a minimum at the ultimate.

In the transverse test, with a good load deflection, we find that the product of the load times deflection gives a close approximation of the impact resistance. The figure is as accurate as the shock test itself.

W. J. GREDE³: In going over Mr. Boegehold's paper, I am not sure that I understood Mr. Harrington's interpretation of the data on page 582 in Klingenstein's drill test, under (c) "Chromium increase from 0.3 to 0.7 per cent, decreased wear resistance 25 per cent." I am wondering if Mr. Harrington misquoted. He said that it decreased resistance to wear. In other words, wear is decreased 25 per cent by the addition of chromium. Is that your interpretation?

R. F. HARRINGTON⁴: Between 0.3 and 0.7 per cent, wear is reduced

³ Liberty Foundry Inc., Wauwautosa, Wis.

⁴ Hunt-Spiller Mfg. Corp., Boston.

or resistance to wear is increased. I have been interested vitally in the question of wear. When Mr. MacPherran and I were in Europe in 1929, we made a particular effort to get the reaction of various engineers in European plants and foundries on this subject of wear, also the methods of measuring the wear or wear resistance of an iron. I remember in a large diesel engine plant they told us that they had found, after a great many investigations, that the only true method for determining wear was placing the material actually in the engine.

They pointed out that to test an iron without lubrication when it was to operate with lubrication, gave absolutely no indication of the wear resistance that might be expected of the iron under operating conditions.

We believe that when an iron operates in a cylinder with a minimum amount of lubrication, we may desire an entirely different type of iron than when it operates with a large amount of lubrication, as in the case of an automotive cylinder.

The question of high versus low phosphorus in iron was discussed with many European metallurgists. It was interesting to have them tell us that in their judgment the phosphorus should be run from 0.50 to 0.1 per cent. The limiting factor in that connection was the fact that the casting might fail because of low resistance to impact or low resistance as far as other stresses were concerned.

Since 1929, it seems that there is a slight tendency to get away from the idea that irons between 0.50 and 0.10 per cent phosphorus show greater wear resistance than those on the lower border of phosphorus. One large manufacturer of diesel castings now has adopted the low phosphorus idea, in the sense that they now have demonstrated that the lower phosphorus irons have given as good wear resistance as the higher phosphorus irons.

J. W. BOLTON⁶: Wear resistance is an extremely complex property and, in materials in general, three factors, possibly more, are involved. One is the resistance to attrition or wearing away. Another is the resistance to galling or seizing. The third is resistance to different types of chemical decomposition. As a result of the cumulation of these and other factors, each time conditions are changed, an entirely new picture is presented.

For example, in one type of wear, where the intensity of pressure is comparatively low, the danger of galling, seizing or breaking off the structure is negligible; in others, such as in the valve industry, the intensity of pressure may be high and we are interested in seeing that the material does not seize or gall.

In seizing or galling, microscopic examination shows that there is practically a welding action between the two pieces if the temperatures are increased sufficiently.

Take the question of high and low phosphorus irons. The hard steadite forms a fairly good riding or bearing surface if the pressure intensity is not too high. There is little attrition and practically no tendency toward galling or seizing. However, if the pressure is increased

⁶ Lukenheimer Co., Cincinnati.

sufficiently, surface distortions result and break off the hard steadite particles. Those act as abrasives and rapidly chew up the material.

There is one property of cast iron that should be extremely interesting to producers. It is something he can use in his sales propaganda. That is the unusual freedom of gray iron from the tendency to seize or gall.

Some tests were presented before the A.S.T.M. in 1928 by N. L. Mochel,* Westinghouse Electric & Mfg. Co., Philadelphia, where he showed the tendencies toward seizing of certain metals, gray iron, alloys and nitrided materials. Others have repeated these experiments in various manners. Mr. Mochel was running under light pressures. We have done work under quite heavy pressures.

An interesting point appears in that connection. Cast iron usually breaks down by reason of its mechanical weakness. Little particles are torn off. The more brittle these particles are, as in the case of steadite, the more likely they are to be torn off by sufficiently high pressure intensity. There is practically no tendency to gall, seize or weld together. If two pieces of stainless steel are heated to 750 degrees Fahr., and are rubbed together slowly under fairly high pressure, they will almost weld together. They seize upon one another. Cast iron is practically free from that difficulty.

Another extremely interesting test confirmed the old-timers' claim that graphite in cast iron acts as a self lubricant. If the pressure is high enough with a temperature of 750 degrees Fahr., the graphite comes out and forms a film on the surface.

J. T. MACKENZIE: There is one word that I want to speak for the transverse test. That is with respect to the newly popularized feature of the damping capacity of cast iron in connection with vibration absorption. In damping capacity, cast iron is far ahead of the field. The damping capacity is inversely proportional to the modulus of elasticity. It is proportional to the deflection. An iron with 0.07 in. deflection, will show seven-fifths as good damping capacity as one with 0.05 in. deflection. Thus, in making machinery to withstand vibrations, the deflection is the most satisfactory test.

W. W. KERLIN: I would like to confirm what Mr. MacKenzie has said. Recently, we made tests on the damping capacity of gray iron. Taking steel at 100 as the best, we found that a Class 20 iron when cast in a $\frac{3}{8}$ -inch section, had a damping capacity of about 300, whereas a Class 30 iron, cast in the same section, had a damping capacity of about 200. It seems as though the higher tensile irons are less efficient in absorbing vibration than the lower tensile irons. That checks with the transverse rigidity test. With a given load, the higher tensile irons deflect less than the lower tensile irons. Therefore, the modulus of elasticity is higher in proportion.

* *The Seizing of Metals at High Temperatures, Proceedings, A.S.T.M., Vol. 28, pp. 269-277, 1928.*

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